MAR 12 1929

BUREAU OF STANDARDS JOURNAL OF RESEARCH

Vol. 1 SEPTEMBER, 1928 No. 3

CONTENTS

RP. No.	TITLE AND AUTHOR	Page
11	Standard Solar Wave Lengths (3592–7148A). Keivin Burns, William F. Meggers, and C. C. Kiess	297
12	Wave Lengths and Zeeman Effects in Yttrium Spectra. William F. Meggers	319
13	Wear and Mechanical Properties of Railroad Bearing Bronzes at Different Temperatures. H. J. French, S. J. Rosenberg, W. Le C. Harbaugh, and H. C. Cross	343
14	Steel For Casehardening—Normal and Abnormal Steel. S. Epstein and H. S. Rawdon	423
15	Strain Markings in Mild Steel under Tension. H. S. Rawdon	467

DEPARTMENT OF COMMERCE
BUREAU OF STANDARDS
George K. Burgess, Director

BUREAU OF STANDARDS JOURNAL OF RESEARCH

September, 1928
Vol. 1, No. 3

PRICE, 25 CENTS
$2.75 PER YEAR ON SUBSCRIPTION
Sold only by the Superintendent of Documents, U. S. Government Printing Office
Washington, D. C.

UNITED STATES
GOVERNMENT PRINTING OFFICE
WASHINGTON
1928

BUREAU OF STANDARDS JOURNAL OF RESEARCH

NOTICE

Beginning July, 1928, a new monthly periodical, the BUREAU OF STANDARDS JOURNAL OF RESEARCH, continued the publication of the two series of research papers heretofore issued as "Scientific Papers" and "Technologic Papers." Forty-four volumes (22 of each of the two superseded series) have been published, comprising some 942 research papers (572 on fundamental science and 370 on applied science).

The new journal contains the bureau's research papers and critical reviews in the fields of science and technology. These will be comparable in interest and importance with the bureau papers already issued. The union of pure and applied science in one journal will, it is believed, tend to bridge the gap between the two fields, and by so much shorten the lag between discovery and its application. This makes it the more desirable that all engaged in scientific or technical work should have available for current use and permanent reference the new BUREAU OF STANDARDS JOURNAL OF RESEARCH. Each volume (semiannual) will be indexed and a cumulative consolidated index will be included in the bureau's list of publications as heretofore. It is expected the monthly issues will range from 100 to 300 pages, an average issue containing about 200 pages. Subscriptions ($2.75 per year, United States, Canada, and Mexico) should be placed direct with the Superintendent of Documents, U. S. Government Printing Office, Washington, D. C. ($3.50 for foreign subscriptions). Separates of the articles printed in the JOURNAL may be purchased from the Superintendent of Documents.

STANDARD SOLAR WAVE LENGTHS (3592–7148A)

By Keivin Burns, William F. Meggers, and C. C. Kiess

ABSTRACT

In order to determine a series of solar wave lengths in the international system to serve as standards for a new table of solar wave lengths and to find the exact corrections which should be applied to Rowland's classical table of solar spectrum wave lengths, the Bureau of Standards and the Allegheny Observatory cooperated in measuring the wave lengths corresponding to selected dark (Fraunhofer) lines in the solar spectrum. The wave lengths were compared with those of standard neon lines by the Fabry and Perot étalon-interferometer method, both spectra being photographed simultaneously. More than 11,000 observations were made on 729 solar lines in the wave-length interval (octave) from 3592.027A in the ultra-violet to 7148.159A in the red. The majority of these standard solar wave lengths have individual probable errors smaller than 1 part in 4,500,000.

CONTENTS

	Page
I. Introduction	297
II. Method of observing	299
III. Description of apparatus	300
IV. Comparison sources	302
V. Measurement and reduction	303
VI. Results	304

I. INTRODUCTION

The classical spectroscopic investigations of Professor Rowland culminating in his Preliminary Table of Solar Spectrum Wave Lengths,[1] formed the foundation of the so-called Rowland scale of wave lengths which has been in more or less universal use for the past 35 years. Beginning with the best available value of the wave lengths corresponding to the sodium yellow lines,[2] the relative wave lengths of a selected list of solar absorption lines[3] were determined by the method of overlapping spectra from diffraction gratings.

Early in the present century it was recognized that Rowland's standards were too large by approximately 1 part in 30,000, and relative errors of the order of 1 part in 100,000 were detected in different spectral regions. The International Union for Cooperation in Solar Research,[4] later replaced by the International Astronomical

[1] Rowland, Preliminary Table of Solar Spectrum Wave Lengths, The Univ. of Chicago Press; 1896.
[2] Rowland, Phil. Mag. (5), **23**, p. 257; 1887 and **27**, p. 479; 1889.
[3] Rowland, Phil. Mag. (5), **36**, p. 49; 1893.
[4] International Union for Cooperation in Solar Research, Trans. I, p. 230; 1906; II, pp. 109–174; 1908; III p. 139; 1911; IV, p. 59; 1914.

Union,[5] has dealt with the situation since 1905, and as a result of recommendations a new set of standard wave lengths has been determined and adopted to form the basis of the so-called international scale. This international system consists (1) of a primary standard—viz, the wave length of cadmium red radiation which had previously been compared with the meter, and (2) several hundred secondary standards (3370–6750A) derived from the emission spectrum of the iron arc. These secondary standards were measured relative to the primary standard by means of the Fabry and Perot[6] étalon-interferometer method. Certain neon lines which have been compared with cadmium by the same method have also been adopted as secondary standards.

This international system has gradually been displacing the older Rowland system, but there has been a tendency, especially in astrophysics, to retain the latter. The solar spectrum tables of Rowland are an almost inexhaustible mine of astrophysical study, but they could not be translated to the international scale without accurately determined corrections. On the other hand, the use of two different systems of standards caused considerable confusion in spectroscopy, and the only remedy for this double standard appeared to lie in a redetermination of solar spectrum wave lengths on the international scale. In 1922 the International Astronomical Union expressed itself as follows:[7]

> It is believed that the time has arrived for the determination of a series of solar wave lengths in the international system to serve as standards for a new table of solar wave lengths. It is recommended, therefore, that efforts be made to obtain at least three independent determinations of the wave lengths of a selected list of solar lines.

In 1923 the spectroscopy section of this bureau united with the Allegheny Observatory in a cooperative program of solar wavelength determinations, and results for one octave of the spectrum (3592–7148A) were completed in 1927. These results were obtained in four overlapping spectral intervals. They were published in three installments according as the work in different spectral regions reached completion.[8] In order that these results may be presented as a unified whole, they are republished here with the wave lengths (Table 1) in consecutive order and mean values replacing those reported twice in the slightly overlapping regions covered by the first publications. Similarly, the corrections to Rowland's wave

[5] International Astronomical Union, Trans. I, p. 35; 1922; II, p. 40; 1925.

[6] Fabry and Perot, Ann de Chim, et de Phys., **25**, p. 98; 1902; Astrophys. J., **15**, pp. 73 and 261; 1902.

[7] I. A. U. Transactions I, p. 36; 1922.

[8] Burns and Meggers, Standard solar wave lengths (4073—4754A), Pub. Allegheny Observatory, **6**, No. 7; 1925. Burns and Kiess, Standard solar wave lengths (5805–7142A), Pub. Allegheny Observatory, **6**, No. 8; 1927. Burns, Standard solar wave lengths (3592–4107A and 4761–5892A) Pub. Allegheny Observatory, **6**, No. 9; 1927.

lengths, previously reported in four sections, are now collected in one table (Table 2). A large number of additional "sun-arc" observations are included in Table 1, but the discussion of these is reserved for another paper. The main purpose of the one in hand is to present in complete and final form our observations on standard solar wave lengths. Only the essential facts relating to the method of observing, the description of the apparatus, and of the comparison sources will be given here. Further details may be found in the earlier publications.

II. METHOD OF OBSERVING

The method of observing is essentially that devised and first used by Fabry and Perot [9] in 1902 for the measurement of 33 solar wave lengths (4643-6471A). It involves the use of a Fabry and Perot étalon interferometer combined with a prismatic or grating dispersing system, so that the orders of interference at various points in the spectrum may be compared. The method has become more or less familiar on account of its repeated use for the establishment of the international system of secondary standards in the emission spectrum of the iron arc. In the case of an absorption spectrum like that of the sun interference patterns similar to those produced by bright lines are obtained for dark (Fraunhofer) lines by a choice of order of interference and slit width such that a practical disappearance of the channeled spectrum from the continuous background is accomplished and interference fringes appear on the dark lines. The theory of interferences produced by the dark lines of the solar spectrum was discussed in some detail by Fabry and Buisson [10] in 1910, but no further measurements of solar wave lengths were published until ours appeared. Similar observations have recently been made by this method at the Mount Wilson Observatory, and a portion of the results have been published. [11]

In order that solar spectrum standards may be placed on the same footing as the international secondary standards, they must be compared with the primary standard or its equivalent. Recognizing certain advantages of neon lines for such comparisons the International Astronomical Union [12] considers the mean of eight or more well-determined neon wave lengths as practically equivalent to the primary standard. Our method of observing was arranged so that the solar spectrum and neon comparison spectrum could be recorded simultaneously. This procedure eliminates, or reduces to a minimum, the small uncontrollable disturbances which otherwise are likely to introduce systematic errors in precise comparisons of wave lengths.

[9] Fabry and Perot, see reference No. 6.
[10] Fabry and Buisson, J. de Phys., 9, p. 197; 1910.
[11] Babcock, Astrophys. J., 65, p. 140, 1927.
[12] I. A. U. Transactions II, p. 47; 1925.

III. DESCRIPTION OF APPARATUS

All of the spectrograms were made at the Allegheny Observatory where the coelostat and spectrograph described by Schlesinger[13] were employed, but the location and mounting of the latter instrument was changed. The former vertical mounting of the spectrograph proved to be too unstable for the rather long exposures required in this work. The instrument was, therefore, mounted horizontally on concrete piers in the basement of the observatory, where ideal stability and freedom from temperature changes were obtained. A coelostat mirror reflected the sunlight down the polar axis to another mirror which sent the light into the solar telescope and interferometer. The optical arrangement from this point onward is shown diagrammatically in Figure 1. After leaving the second coelostat mirror, the beam of sunlight passed through a color filter which was placed immediately in front of the solar lens. This filter reduced the heating effect of the sun in the interferometer, and at the same time removed the part of the solar spectrum corresponding to the neon comparison source, so that the spectrum of the latter could be simultaneously recorded in an auxiliary spectrograph camera. In the interval 5,800–6,500A the two spectra were photographed superposed, higher orders than the second being removed by a yellow-glass filter.

A telescope of 40 cm focal length forms an image of the sun in the interferometer, which is of the Fabry-Perot type. A diaphragm 6 mm in diameter near the first plate of the interferometer restricts the beam of the comparison spectrum to the size of the solar image. The sunlight from the telescope passes through a diagonal mirror, which serves to reflect the light of the laterally situated neon comparison lamp into the path of the solar beam. From this point on the solar and comparison beams follow identical paths until the grating separates them into spectra. For observing the photographic region of the solar spectrum this diagonal mirror is lightly silvered, reflecting rather more light than it transmits. For the longer solar wave lengths a thinner coat of silver is used, and for the region longer than 6600A a clear-glass mirror has been used for some of the observations.

The interferometer is so mounted on a stand as to permit the necessary adjustments or to allow it to be lowered out of the path of the light so that the centering of the solar image and the comparison source on the grating may be more readily effected. The interferometer is turned 90°, in order to adjust the parallelism of the plates by means of a mercury lamp. The interferometer plates are thinly coated with silver, deposited cathodically. Some 10 to 15 images of a 40-watt lamp can be seen through a pair of plates. While a thicker

[13] Schlesinger, Pub. Allegheny Obs., 3, p. 99; 1914.

coat would give slightly better interference, the increase in exposure time necessitated by the use of dense films more than offsets the slight gain in resolving power. The separators, or étalons, are made of invar.

A telescope of 25 cm focal length projects the ring system upon the slit of the spectrograph. The objective, of quartz and fluorite, is achromatic for the whole region of the solar spectrum observable at Allegheny Observatory. The interferometer is set at such a distance from this objective that the image of the diaphragm, 6 mm in diameter, is slightly smaller than the grating.

A gauge plate, or repère, having five fine parallel slots extending at right angles to the slit, is mounted on ways which allow it to slide almost in contact with the latter. Before the interferometer is placed in position this repère plate is slid in front of the slit. The intersections of the slit and the slots in this plate give five point-sources

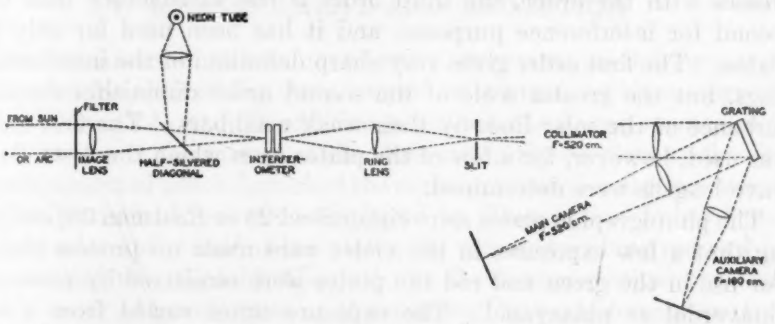

FIG. 1.—*Diagram of interferometer and grating spectrograph for observing solar wave lengths*

of light whose images are photographed in both the solar and comparison spectra. This furnishes a means of making sure that the same scale is used in computing the ring diameters in solar and comparison spectra, a matter of considerable importance when the two spectra differ greatly in wave length, but of little consequence when both spectra are of nearly the same wave length.

The collimator and camera lenses of the spectroscope are of 19 cm aperture and 520 cm focal length. Since parallel light falls upon the grating and the rulings are only 10 cm long, only the central 10 cm of each lens is useful in connection with the interferometer. The lenses are corrected for the photographic region, having minimum focal length at 4500A, at which point in the spectrum the plate is perpendicular to the axis of the camera. The grating, by Michelson, is ruled with 500 lines per millimeter, giving a linear scale of 3.65A per mm in the first order.

To photograph the comparison spectrum, an auxiliary camera has been installed as shown in Figure 1. The lens of this camera is of 10 cm aperture and 160 cm focal length. The field is good over

an angle of 5°, which is more than is required to photograph the neon spectrum in the region 5852 to 6598A. Since the solar camera is fixed with respect to the collimator, it is necessary to move the auxiliary camera as the various regions of the spectrum are brought into the solar camera. This comparison camera is arranged to rotate around an axis passing through the center of the grating; consequently the focus for the neon spectrum remains constant as the camera is rotated. For the region shorter than 6600A the comparison spectrum is observed in the first order on the side of zero opposite from that on which the solar spectrum is photographed. For the longer region it is preferable to observe the comparison and solar spectra in the first and second orders, respectively, on the same side of zero.

The greater portion of the observing has been done with the grating adjusted to photograph the solar spectrum in the second order, giving a scale of 1.82A per mm. Because of a slight astigmatism that increases with the order, the third order is less satisfactory than the second for interference purposes, and it has been used for only 10 plates. The first order gives very sharp definition of the interference rings, but the greater scale of the second order diminishes the disturbance of the solar lines by their weak neighbors. The first order was used, however, for a few of the plates from which the A. O. B. S. wave lengths were determined.

The photographic plates were either Seed 23 or Eastman 33, excepting that a few exposures in the violet were made on process plates. For use in the green and red the plates were sensitized by means of pinaverdol or pinacyanol. The exposure times varied from a few minutes to over two hours, depending on the condition of the sky, the state of the various silver coats, and the slit width.

Spectrograms were made in each portion of the solar spectrum with a variety of étalons and with two or more sets of interferometer plates. The various étalons were made of invar and had lengths of 3.75, 5, 6, 7.5, 8, 10, 12, 15, and 20 mm.

IV. COMPARISON SOURCES

The wave lengths of Table 1 were derived by simultaneous comparison with the red lines of neon, as already stated. The neon source was a lamp of the type described by Nutting.[14] Its electrodes are aluminum disks 25 mm in diameter; the capillary is 10 cm long and has a bore of 2.5 mm. An exciting current of about 50 m. a. was delivered by 5,000 v. a. c. without capacity. The tube was observed exactly side on. The particular tube used in this investigation had been used previously in a comparison of neon wave lengths with the primary cadmium standard.

[14] Nutting, Bull. Bureau of Standards, 4, p. 511; 1908.

The maximum pressure in the solar reversing layer where the Fraunhofer lines originate has been demonstrated to be so low that for spectroscopic purposes it may be regarded comparable with laboratory vacuum sources. For comparison of solar with terrestrial wave lengths we have employed the very convenient type of vacuum arc designed and built by Curtis.[15] Using currents of 6 amp. or less, this arc operates without a cooling system and is in other respects so simple to manipulate that it would seem that this or a similar type of source should become a standard piece of laboratory equipment wherever wave lengths of a high order of accuracy are being determined. This arc was used as the source for all the sun-arc comparisons listed in Table 1. It was usually operated at pressures corresponding to 1 to 6 cm of mercury. All the vacuum arc wave lengths were derived from the neon standards by simultaneous comparison exactly in the same manner as the solar wave lengths.

V. MEASUREMENT AND REDUCTION

The diameters of two or, more often, three innermost fringes were measured for each of the wave lengths on a number of spectrograms ranging from 2 to 33. In addition to the solar and neon spectra, each plate had on it the images of the slots in the repére plate at the slit, measurement of which furnished the scale factor used in the reductions.

The method of reduction employed by Buisson and Fabry [16] was followed for about half of the plates, the remainder being reduced by the method described by Childs.[17] In the former method the fractional order of interference at the center of the interference pattern is derived from measurements of the repère images and ring diameters, while in the latter it is deduced from ring diameters alone when three or more are available. Practically the same results are obtained from both methods, but the latter has the advantage of involving considerably less labor.

The order of interference was invariably obtained with the aid of the neon interferences,[18] and the thickness of the interferometer, or double separation of the plates, was computed from each of six or more neon lines, the mean optical distance thus found being used to derive the solar wave lengths from their measured orders of interference. The values used for the neon lines were those compared with the primary cadmium standard; [19] their mean value is regarded as practically equivalent to the primary standard itself.

Corrections for dispersion of phase upon reflection in the interferometer surfaces were made by the well-known device of comparing

[15] Curtis, J. Opt. Soc. Am., **8**, p. 697; 1924.
[16] Buisson and Fabry, J. de Phys., **7**, p. 169; 1908, Astrophys. J., **28**, p. 169; 1908.
[17] Childs, J. Sci. Inst., **3**, pp. 97 and 129; 1926.
[18] Meggers, Bull. Bureau of Standards, **12**, p. 203; 1915.
[19] Burns, Meggers, and Merrill, Bull. Bureau of Standards, **14**, p. 765; 1918. I. A. U. Trans. I., p. 40, 1922. II, p. 44, 1925.

wave lengths first with low orders of interference and then with high values in which the differential phase change plays a proportionately smaller part.[20] For this purpose wave lengths of the iron arc in vacuo were compared with neon standards by means of interferometers of 3.75, 6, 10, and 20 mm separation.

Observations made under conditions of temperature and pressure of air which deviated from normal were corrected to their values at 15° C. and 760 mm Hg by the tables of Meggers and Peters.[21]

The proper corrections for motion of the earth relative to the sun were made for each spectrogram.

VI. RESULTS

The first column of Table 1 contains the wave lengths of solar lines compared with red neon standards by means of simultaneous exposures as described above. The second column indicates the probable error of measurement. These probable errors are expressed in parts per million; thus "A" denotes a probable error of 1 part in 6,000,000; "B," 1 part in 4,500,000; "C," 1 part in 3,000,000; and "D" probable errors greater than 1 part 3,000,000. The number of independent comparisons of each line with the neon standards is indicated in column 3. The fourth, fifth, and sixth columns contain Rowland's wave length, chemical identification, and intensity for each line. An "s" attached to Rowland's wave length means that it was one of his standards. Column 7 shows the wave-length difference between Rowland and A. O. B. S. for the individual lines, and the next column gives the value of this difference interpolated from Table 2. In the ninth column are shown the differences between the wave lengths in the solar spectrum and the laboratory vacuum-arc values for the same lines, the unit being 0.001A. The final column gives the temperature class according to King and the pressure class to which the line belongs. The A. O. B. S. wave-length values are for air at 15° C. and 760 mm Hg pressure, while those of Rowland apply to air at 20° C. and normal pressure. In both cases the source was integrated solar light.

In Table 2 are listed the corrections which must be applied to Rowland's wave lengths to reduce them to the international scale. These are taken from a smooth curve drawn through the plotted values of column 7 in Table 1. The differences, Rowland minus A. O. B. S., run smoothly for the most part, but there are several marked discontinuities, notably at 4142A, at 5220A, and at 6870A. The smaller irregularity near 5800A may not be real, as the observations between 5740 and 5850A are of slightly inferior quality.

[20] Meggers, Bull. Bureau of Standards, **12**, p. 199; 1915.
[21] Meggers and Peters, Bull. Bureau of Standards, **14**, p. 724; 1918.

A most encouraging feature of this comparison is an entire lack of an intensity equation. The values of the differences, Rowland minus A. O. B. S., are quite independent of the intensity of the solar line. After applying the interpolated corrections to Rowland's values, the mean of the residuals, Rowland minus A. O. B. S., is only ± 0.0015A.

TABLE 1.—*A. O. B. S. solar wave lengths*

[3592-7148A]

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
3592.027	C	3	.169	V?	2	142	141		VE
3600.736	C	3	.880	V	3	144	141		
15.661	C	3	.802	Fe	3	141	141		
35.470	C	3	.608 s	Ti, Fe	4	138	141		
50.539	C	3	.681		2	142	141		
61.366	C	6	.506	Fe	3	140	140		
72.712	C	6	.851	Fe	3	139	140		IV
81.648	B	8	.787	Fe	3	139	140		IV
3695.056	B	6	.194 s	Fe	5	138	140	+6	IV
3710.291	A	10	.431	Y	3	140	140		
25.495	A	10	.638		3	143	140		
41.064	A	10	.205	Ti	4	141	140	+5	I
52.419	A	7	.556	Fe	3	137	140		
60.537	A	10	.679	Fe	4	142	140		III
69.993	A	9	0.132 s	Fe	4	139	140	+6	IV
81.190	A	10	.330 s	Fe	3	140	140		IV
3793.877	A	10	4.016 s	Fe	2	139	140		IV
3802.285	B	10	.424	Fe	2	139	140		
04.014	A	9	.151 s	Fe	3	137	140		
21.187	A	9	.328 s	Fe	4	141	140		IV
36.090	A	10	.229 s		2	139	140		
43.264	B	9	.404 s	Fe	4	139	140	+7	IV
59.219	C	8	.355	Fe	3	136	140		III
64.302	B	8	.438 s	C	3	136	140		
73.764	B	8	.903	Fe	4	139	140	+3	IV
85.516	A	14	.657	Fe	4	141	140	+2	III b
3897.458	A	14	.596 s	Fe	2	138	140		IV
3906.751	A	17	.890	Fe	4	139	140		
16.736	A	18	.879 s	Fe	5	141	140	+5	IV b
37.336	A	9	.479 s	Fe	3	143	140	+8	IV b
49.958	A	18	0.102 s	Fe	5	144	141	+5	III b
53.859	A	19	4.002 s	Fe	3	143	142		
58.739	B	6	.877	Fe	1	138	142		
60.283	A	17	.422 s	Fe	4	139	142		
63.690	A	14	.831	Cr	3	141	143		
76.866	B	8	7.009	Fe	2	143	145		
77.747	C	16	.891	Fe	6	144	145	+6	III b4
78.658	A	9	.809	Co, Cr	3	151	145		
81.770	A	11	.917	Ti	4	147	146	+9	II
87.609	A	9	.755	Ti?	2	146	147		
91.120	A	13	.333	Cr, Zn	3		147		
3993.099	A	8	.246	Fe	2	147	148		
4000.255	C	5	.403	Fe	2	150	149	±0	
00.460	C	6	.611	Fe	2	151	149		b
03.768	A	11	.912 s	Co, Fe, Ti	3	144	150		
10.174	C	5	.327	Fe	1	153	151		
11.713	B	8	.865	Fe	2	152	151		
16.421	A	11	.574	Fe	2	153	151		
22.736	A	8	.893	Fe	2	156	152		
29.639	A	10	.796	Fe, Zn	5	157	154		
4030.187	A	12	.339	Fe	2	152	154		
31.962	A	6	2.117	Fe	2	155	154		
37.119	B	5	.268		2	149	154		
47.306	A	7	.461		2	155	155		
49.327	B	7	.482	Fe	2	155	155	−1	

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
4050.675	A	8	.830	Fe	2	155	155		
53.824	A	12	.981	Fe, Ti	3	157	155		
59.718	A	8	.872	Fe	2	154	155		
62.446	A	11	.599 s	Fe	5	153	155	+6	III b
66.585	A	6	.742	Fe	2	157	155		
73.766	A	31	.921 s	Fe	4	155	155	+6	
79.841	A	41	.966	Fe	3	155	155	+3	IV b
80.879	A	35	1.033	Fe	2	154	155		
82.941	A	25	3.095	Mn	4	154	155	+4	
84.499	A	25	.647	Fe	5	148	155	+8	IV d
85.007	A	28	.161	Fe	4	154	155		IV b
91.556	A	38	.711	Fe	3	155	156		
94.938	A	22	5.094	Ca?	4	156	156		
4098.179	B	26	.335	Fe	5	156	156	+3	IV b
4100.744	A	32	.901	Fe	4	157	156	+5	IIA
01.267	A	10	.421	Fe	2	154	156		d
06.290 [1]	A	10	.420	Fe	2	160	156	±0	
06.431 [1]	B	10	.583	Fe	2	152	156	+9	
07.492	A	25	.649	Ce-Fe-Zr	5	157	156	+4	III b
12.321	B	20	.478	Fe	2	157	156		
14.450	B	17	.606 s	Fe	4	156	156	+5	IV b
20.212	C	13	.368	Fe	4	156	155	+6	IV b
26.190	B	17	.344	Fe	4	154	154	+7	IV b
36.527	A	20	.678	Fe	4	151	152		d
37.000	C	18	.156	Fe	6	156	152	+3	IV b
39.936	A	23	0.089	Fe	6	153	152	+9	IIA a
43.876	C	6	4.038	Fe	15	162	163	+9	b1
47.675	C	13	.836	Fe	4	161	163	+6	III b
49.371	A	7	.533	Fe	4	162	163	+6	V e
54.505	C	12	.667	Fe	4	162	164	+6	III b
54.814	A	17	.976	Fe	4	162	164	+8	IV d
58.794	B	19	.959	Fe	5	165	164	+1	V d
63.651	C	11	.818	Ti, Cr —	4	167	164		
68.619	C	8	.784		2	165	163		
78.858	B	13	9.025		3	167	162		
82.387	A	25	.548	Fe	3	161	162		IV b
84.898	B	12	5.058 s	Fe, Cr	4	160	162	+6	III b
87.046	C	8	.204	Fe	6	158	162	+8	III d
91.683	C	11	.843	Fe	3	160	161		e?
4198.639	C	16	.800	Fe	3	161	161		V d
4202.041	C	8	.198	Fe	8	157	160	+12	I b1
07.132	B	26	.291	Fe	3	159	160		IV b
08.608	B	20	.766	Fe	3	158	160	+3	V b
13.651	B	26	.812	Fe	3	161	160	+4	IV b
20.348	B	30	.509	Fe	3	161	159	+8	IV b
22.219	B	22	.382 s	Fe	5	163	159	+7	III d
24.175	C	9	.337	Fe	4	162	159	+4	IV e
25.461	C	16	.619	Fe	3	158	159	+7	IV d
25.961	B	16	6.116	Fe	2	155	159		IV b
26.742	C	6	.904 s g	Ca	20d?	160	159	+13	I
27.432 [2]	C	9	.606	Fe	4	174	159	+6	III d
31.023	C	11	.183	Ni	4N	160	159		
32.731	B	17	.887	Fe	2	156	159	+6	IA
33.612	C	13	.772	Fe	6	160	159	+10	III d5
35.947	C	10	6.112	Fe	8	165	159	+10	III d5
4238.025	D	5	.188	Fe	3	163	159	+3	IV d?
41.121	B	25	.285	Fe-Zr	2	164	159		
43.815 [3]	D	4	.981		2	166	159		
46.838	B	10	.996	Y?	5	158	158		
4248.233 [4]	B	11	.384	Fe	2	151	158		IV b

[1] 4106.2 and 4106.4. These lines are too close together for the most accurate measurement with moderate resolution. The mean wave length of these two lines is probably quite correct.

[2] 4227.4. The A. O. B. S. and laboratory values of this wave length agree. This is a line of much greater solar intensity than 4, the figure in Rowland's table. Perhaps there is a typographical error in Rowland's value of the wave length.

[3] 4243.8. A close companion renders the interference measurements of this line rather uncertain.

[4] 4248.2. The laboratory value of this line agrees with Rowland.

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A.O.B.S.λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
4250.129	C	7	.287	Fe	8	158	158	+10	III c5
54.347	C	5	.505 s	Cr	8	158	158	+10	II
57.660	A	11	.815	Mn	2	155	158		II
60.488	C	7	.640 s	Fe	10	152	158	+14	III c2
65.264	B	20	.418	Fe	2	154	158		
66.967	C	11	7.122	Fe	3	155	158	+2	IV b
68.754	A	17	.915	Fe	2	161	158	+10	IV b
71.160	C	7	.325	Fe	6	165	158	+7	III d
71.774	C	7	.934 s	Fe	15	160	158	+14	II b1
76.678	B	13	.836	Zr	2	158	158		
82.411	B	19	.565	Fe	5	154	158	+8	III b1
90.384	C	7	.542	Fe	1	158	158		
4291.471	B	26	.630	Fe	2	159	158	+7	IA a3
4318.658	B	12	.817	Ca, Mn?	4	159	160		
21.797	B	9	.961	Fe	2	164	161		
25.773	D	5	.939 s	Fe	8	166	161	+12	II b1
26.758	B	14	.923	Fe	2	165	161		
27.110	A	18	.274	Fe	3	164	161		V b
31.649	B	18	.811	Ni	2	162	162		II
37.055	C	20	.216	Fe	5	161	162	+9	II b3
37.927	A	17	8.084	Ti	4	157	162	+11	VE
38.265	C	17	.430	Fe	1	165	162	+2	
43.703	B	18	.861	Fe	2	158	162		
46.557	C	13	.725	Fe	2	168	162	+4	
47.238	B	17	.403	Fe	1	165	162	+5	
48.946	B	16	9.107	Fe	2	161	162	+10	
51.548	B	12	.711	Fe	2	163	162	+6	IV b
58.504	B	13	.670	Fe	2	166	161	+4	IV b
59.621	D	6	.748 s	Cr	3	163	161		I
60.799	A	11	.958	Fe, Zr	1	159	161		
65.903	B	28	6.061	Fe	2	158	161		
67.910	A	24	8.071	Fe	2	161	161	+4	IIIA a
69.779	C	16	.941 s	Fe	4	162	161	+7	III b3
73.564	B	18	.727	Fe	2	163	161	+1	
75.947	C	25	6.107 s	Fe	6	160	160	+17	I a3
77.703	B	18	.948	Fe	1	155	160		
79.241	D	5	.396	V	4	155	159		
83.563	C	9	.720 s	Fe	15	157	159	+19	II b1
87.897	C	22	8.057	Fe, Co	2	160	159		IV b
88.414	C	7	.571	Fe	3	157	159	+7	IV d?
89.252	A	18	.413	Fe—	2	161	159		IIA a
94.066	A	10	.225	Ti?	2	159	159		
98.020	B	16	.178		1	158	150		
4399.775	A	15	.935	Ti, Cr	3	160	160		
4404.767	C	9	.927 s	Fe	10	160	160	+17	II b1
08.204	B	14	.364	V	2	160	160		
15.144 [1]	B	13	.263 s	Fe	8	149	161	+22	II b1
16.829	B	17	.985		2	156	161		
22.572	B	16	.741	Fe, Y	3	169	161	+3	III b3
4425.445	B	19	.608 s	Ca	4	163	161		I
27.323	B	16	.482	Fe	5	159	162	+13	I a3
30.195	D	6	.356	Fe	1	161	162		IV
30.621	B	29	.785	Fe	3	164	162	+7	III c4
32.572	B	16	.736	Fe	1	164	162		
33.230	C	16	.390	Fe	3	160	162	+10	IV e
33.783	C	7	.948	Fe	1	165	162	+1	d
35.157	D	5	.321	Fe	2	164	162	+9	IIA a
35.688	B	15	.851 s	Ca	4	163	162		I
39.888	A	24	0.054	Fe	1	166	163		IV
4442.353	B	20	.510	Fe	6	157	163	+13	IV c4

[1] 4415.1. The wave length of this line is no doubt affected by close companions. The laboratory value agrees with Rowland.

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden	Int.	Obs.	Mean		
4443.202	B	13	.365	Fe	3	163	163	+8	III b3
43.812	B	15	.976	Ti	5	164	163	+10	VE
45.476	B	13	.641	Fe	1	165	163	+6	IA
47.737 [6]	A	16	.892 s	Fe	6	155	163	+20	III c4
51.588	B	16	.752	Mn	3	164	163		II
54.388	A	30	.552	Fe	3	164	163	+7	III b3
56.332	B	10	.497	Fe	1	165	163		
59.756	A	15	.922	V	1	166	163		
61.661	B	24	.818	Fe	4	157	163	+8	I a3
66.566	B	22	.727	Fe	5	161	163	+14	II b4
70.136	D	6	.300	Mn	1	164	163		III
70.486	A	12	.648	Ni-Zr	2	162	163		
81.617	A	13	.782	Fe	1	165	163		
84.229	A	17	.392	Fe	4	163	163	+9	IV d
85.682	A	17	.846	Fe	3	164	163	+6	IV e?
91.410	B	16	.570		2	160	164		
94.575	A	14	.738 s	Fe	6	163	164	+12	III c4
4496.864	B	17	7.023 s	Cr	3	159	164		
4501.280	A	16	.448 s	Ti—	5	168	165		
02.220	A	15	.388	Mn	2	168	165		III
04.834	B	16	5.003	Fe	1	169	165	+6	d
08.290	A	23	.455 s	Fe?—	4	165	165		
12.741	A	22	.906	Ti	3	165	165	+7	II
17.536	A	17	.702	Fe	3	166	166	+12	
25.149	A	16	.314	Fe	5	165	167	+11	IV d?
26.934	B	15	7.101	Ca?	3	167	167		
28.631	B	13	.798	Fe	8	167	167	+17	II c4
31.631	A	21	.801	Fe	2	170	167		II b3
34.785	A	17	.953	Ti	4	168	167		
41.524	A	22	.690	Cr	2	166	168		
47.855	A	17	8.024	Fe	3	169	168	+8	b
48.772	A	16	.938	Ti	2	166	168	+8	II
50.775	A	15	.942	Fe?	2	167	168	+5	d
54.042	B	8	.211 s	Ba	8	169	169		II
58.653	A	7	.827	Cr?	3	174	169		
60.094	B	10	.266	Fe	2	172	170	+7	d
63.769	B	20	.939 s	Ti	4	170	171	+8	VE
71.103	A	20	.275 s	Mg	5	172	172		IA
71.983	A	15	.156 s	Ti—	6	173	172		
74.727	B	10	.899	Fe	2	172	172		
76.340	A	15	.512		2	172	172		
78.560	A	17	.732 s	Ca	3	172	172		II
4587.135	B	20	.308	Fe	2	173	172	+7	
88.207	A	16	.381 s		3	174	173		
89.955	A	15	0.126 s		3	171	173		
95.367	A	21	.540	Fe	2	173	173	+9	
4598.127	A	14	.303	Fe	3	176	174	+10	d
4602.008	A	17	.183 s	Fe	3	175	174	+7	
02.951	A	17	3.126	Fe	6	175	174	+10	IB
07.657	A	23	.831	Fe	4	174	174	+11	V
16.133	A	17	.305	Cr	4	172	174		I
17.277	A	16	.452	Ti	3	175	174	+8	II
20.520	A	21	.693	Fe	1	173	174		
25.055	A	19	.227	Fe	5	172	174	+10	d
30.128	A	17	.306	Fe	4	178	174	+8	
35.855	A	15	6.027	Fe	2	172	173	+8	
37.511	C	7	.685 s	Fe	5	174	173	+6	IV d
38.019	A	16	.193 s	Fe	4	174	173	+11	IV b
43.472	A	23	.645 s	Fe	4	173	173	+8	
47.443	C	7	.617	Fe	4	174	173	+9	IV b
48.660	B	10	.835 s	Ni	4	175	173		III
52.170	A	16	.343	Cr	5	173	172		I
56.475	A	22	.644	Ti	3	169	172	+7	I
64.795	B	17	.965	Cr	3	170	172		III
4673.174	C	7	.347	Fe	4	173	173	+11	

[6] 4447.7. One series of measurements gives .730 for the fractional part of this wave length. The results from many other plates are in good agreement and give the value tabulated. A companion affects the wave length as measured by the interferometer.

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
4678.171	B	17	.347 s	Cd	3N	176	173		
78.855	B	19	9.027 s	Fe	6	172	173	+9	V
83.567	A	22	.745 s	Fe	3	178	173	+4	
86.221	A	20	.395 s	Ni	3	174	173		III
4690.146	A	22	.317 s	Fe?	4	173	173	+8	
4700.164	A	15	.337	(Fe)	4	173	174	+7	d
03.006	B	11	.177 s	Mg	10	171	174		V
03.817	C	11	.994 s	Ni	3	177	174		
04.956	A	20	5.131	Fe	4	175	174	+9	
09.718	A	13	.896	Mn	2	178	175		III
15.774	B	13	.946	Ni	4	172	175		II
18.428	B	13	.601	Cr	3	173	175		III
21.001	C	5	.179	Fe?	2	178	175		
22.166	A	22	.342 s	Zn	3	176	176		
28.554	B	11	.732	Fe	4	178	176	+7	IV
33.602	A	19	.779	Fe	4	177	177	+11	IB a
36.785	C	14	.963	Fe	6	178	178	+12	II d
41.537	A	18	.718	Fe	3	181	179	+6	V b
45.808	B	17	.992	Fe	4	184	180	+7	V b
52.431	C	8	.613	Ni	3	182	181		
54.047	C	12	.225 s	Mn	7	178	182		I
4761.534	B	9	.718	Mn	3	184	183		III
66.430	B	9	.621	Mn	4	181	183		III
72.826	C	8	3.007	Fe	4	181	183	+10	III b
79.985	C	10	.169	Co	2	184	184		III
86.820	C	8	7.003	Fe	2	183	185	+14	IV? b
88.766	A	10	.952	Fe	3	186	185	+9	
4789.659	C	11	.849	Fe	3	190	185	+9	V
4802.888	A	10	3.072	Fe	2	184	185	+6	
10.539	B	7	.724 s	Zn	3	185	185		
23.516	B	7	.697 s	Mn	5	181	184	+7	I
24.145	D	8	.325 s	Fe	3	180	184		
32.718	B	9	.905	Fe	3	187	183		
39.551	B	9	.734	Fe	3	183	182	+6	
48.257	D	8	.438		2	181	181		
54.875	A	6	5.059	Fe	1	184	179		
59.751	B	5	.928 s	Fe	4	177	179	+10	III c5
70.824	B	7	.996	Ni, Cr	3	172	177		
82.154	D	9	.336	Fe	3	182	176	+9	d
85.437	D	7	.620	Fe	3	183	175	+7	V
4892.867	B	8	3.030	Fe	1	163	175		
4904.424	C	12	.597		3	173	173		
17.239	A	10	.410	Fe	2	171	172		
24.784	A	12	.956 s	Fe	3	172	171	+11	V b
38.184	A	7	.350	Fe	2	166	171	+10	
38.825	D	7	.997	Fe	4	172	171	+13	IV d
39.697	B	13	.868	Fe	3	171	171	+11	IB a
46.401	B	10	.568	Fe	3	167	171	+14	IV d
50.117	B	13	.291	Fe	2	174	171	+11	d
53.217	C	8	.392	Ni	2	175	171		V?
62.583	B	12	.751	Fe	2	168	171		
66.102	B	12	.270	Fe	4	168	171	+13	V d
67.909	C	14	8.080	Fe	3	171	171	+10	e
73.110	C	12	.281 s	Ti, Fe	4	171	171		V
83.262	B	14	.433	Fe	3	171	171	+14	V d
4994.140	B	16	.316 s	Fe	3	176	171	+10	IB a
5001.876	C	15	2.044	Fe	5	168	171	+12	V d
02.802	A	19	.976	Fe	2	174	171	+10	
14.954	C	14	5.123	Fe	3	169	171	+12	V d
28.135	A	17	.308	Fe	2	173	171	+ 8	
39.968	B	20	0.138	Ti	3	170	171	+ 9	I a
48.448	A	13	.612	Fe	3	164	171		
49.834	B	13	0.008 s	Fe	6	174	171	+15	III a
60.080	A	21	.258 s	Fe	3	178	170		
5067.160	B	20	.336	Fe	3	176	170		

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
5068.779	C	15	.944 s	Fe	5	165	169	+13	V d
74.762	C	15	.932	Fe	5	170	169	+14	V e
76.278	B	15	.450	Fe	3	172	169		
79.241	C	14	.409	Fe	4	168	169		IV b
79.749	B	15	.921	Fe	4	172	169	+ 8	IB a
83.351	A	21	.518 s	Fe	4	167	168	+13	IB a
90.787	A	18	.954 s	Fe	5	167	168	+12	
5099.941	A	21	0.108	Ni	2	167	167		V
5109.661	B	17	.827 s	Fe	2	166	167		
15.401	B	16	.566	Ni	2	165	166		
26.204	C	18	.371 s	Fe, Co	2	167	166		
37.394	B	16	.588	Fe	3	164	165	+12	V d
50.852	C	17	1.020 s	Fe	4	168	165	+12	IB a
59.067	B	19	.231 s	Fe	2	164	165		
73.752	A	21	.917 s	Ti	2	165	164	+10	I
85.911	A	20	6.073	Ti	2	162	163		
5198.723	B	21	.888 s	Fe	3	165	162	+11	IV a
5210.396	A	23	.555 s	Ti	3	169	161	+10	I a
17.403	B	21	.552 s	Fe	3	149		+15	V d
25.537	A	23	.695 s	Fe	2	158	155		IA
42.502	A	22	.658 s	Fe	2	156	159	+12	IV a
53.471	A	24	.633 s	Fe	2	162	162		
63.318	A	23	.486	Fe	4	168	165	+13	V d
73.391	C	14	.558 s	Fe, Cr	2	167	167		
5288.536	A	24	.705 s	Fe	2	169	170		
5300.753	A	22	.929 s	Cr	2	176	172		I
07.370	A	23	.541 s	Fe	3	171	173	+10	III? a
22.051	A	14	.227	Fe	3	176	177		
32.909	A	13	3.089 s	Fe	4	180	179	+8	IB? a4
48.327	A	16	.511	Cr	4	184	183		I
65.409	A	16	.596	Fe	3	187	187		V a
79.583	A	14	.775 s	Fe	3	192	192		a
89.487	A	14	.683 s	Fe	3	196	195	+7	
5398.290	A	14	.486	Fe	3	196	197		
5409.799	A	14	0.000 s	Cr	4	201	201		I
15.215	C	11	.416 s	Fe, V	5	201	202		
5432.956	B	14	3.160	Fe	2	204	203		
45.055	A	16	.259	Fe	4	204	204	+13	V e
62.971	A	15	3.174 s	Fe	3	203	204		e
73.909	A	18	4.113	Fe	3	204	204		d
5487.754	B	19	.959 s	Fe	3	205	205	+ 7	
5501.477	A	18	.683 s	Fe	5	206	206	+11	IB a3
12.989	B	18	3.198 s	Ca	4	209	209		III
25.552	A	17	.765	Fe	2	213	212		
34.848	A	22	5.061 s	Fe	2	213	214		
46.516	B	20	.732	Fe	2	216	216		
60.217	A	24	434	Fe	2	217	216		
76.104	A	24	.320 s	Fe	4	216	217		IV d5
5590.125	A	24	.343 s	Ca	3	218	217		III
5601.286	A	22	.505 s	Ca	3	217	217		III
18.640	A	19	.858	Fe	1	218	218		
24.558	A	20	.769 s	Fe, V	4	211	218		IV
41.447	A	18	.667 s	Fe	2	220	218		
55.499	B	18	.715 s	Fe	2	216	217		V
67.525	B	15	.739	Fe	2	214	216		
79.032	A	17	.249 s	Fe	3	217	216		
5690.433	A	15	.646	Si	3	213	215		
5701.557	A	17	.772 s	Fe	4	215	214	+ 9	III?
17.839	A	16	8.055	Fe	4	216	214		
31.772	A	14	.984 s	Fe	4	212	213		
41.856	C	9	2.068 s	Fe	2	212	212		
52.042	C	13	.254 s	Fe	4	212	211		
60.841	B	10	1.052	Ni	2	211	211		IV
72.156	C	10	.364 s	Si	3	208	210		
5783.871	C	11	4.080 s	Cr	3	209	209		III

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
5797.870	B	9	8.077 s		3	207	209		
5805.226	C	6	.441 s	Ni	4	215	214		V
06.738	D	6	.950 s	Fe	5	212	214		
09.227	B	19	.439 s	Fe	4	212	214		
16.379	C	6	.601 s	Fe	5	222	214	±0	
47.012	D	4	.221	Ni	1	209	213		
48.129	C	14	.342	Fe	3	213	213		
52.235	D	7	.443	Fe	3	208	213		
53.688	A	17	.902 s	Ba	5	214	213	+9	III
55.087	D	4	.300	Fe	1	213	213		
56.101	B	8	.312	Fe	2	211	213		
57.459	B	12	.674 s	Ca	8	215	213	+8	III
57.762	C	6	.976	Ni	3	214	213		IV
59.596	A	25	.809 s	Fe	5	213	213	+3	V
62.368	B	15	.582 s	Fe	6	214	213	+11	V
66.461	A	23	.675	Ti	3	214	213	+8	II
67.573	B	6	.785	Ca	2	212	213		
85.978	D	7	6.193	Air	5	215	215		
87.217	C	4	.445	Air	5	228	215		
91.656	B	4	.878	Air	4	222	215		
92.884	B	19	3.097 s	Ni	4	213	215	+6	II
98.166	B	8	.378 s	Air	4	212	216		
5899.007	C	2	.215	Air	2	208	216		
5901.462	B	12	.682 s	Air	6	220	216		
05.681	A	17	.895 s	Fe	4	214	216	+8	V
08.999	C	2	9.213	Air	3	214	216		
16.257	B	17	.475 s	Fe	3	218	216		
18.441	D	4	.635	Air	4	194	216		
19.052	C	8	.276	Air	5	224	216		
19.645	B	13	.860 s	Air	7	215	216		
24.268	D	5	.490	Air	4	222	216		
27.797	B	15	8.013	Fe	2	216	216		
29.689	C	12	.898	Fe	2	209	216		
30.192	A	16	.406 s	Fe	6	214	216	+11	V
32.092	C	10	.306	Air	5	214	216		
34.665	A	18	.881 s	Fe	5	216	216	+7	V
41.074	C	4	.290	Air	5	216	217		
42.568	C	3	.789	Air	3	221	217		
46.005	C	11	.223	Air	3	218	217		
47.057	D	2	.283	Air	2	226	217		
48.544	B	14	.765 s	Si	6	221	217		
52.726	A	18	.943	Fe	4	217	217		V
53.170	C	6	.386	Ti	1	216	217	+8	II
56.706	B	17	.923 s	Fe	4	217	217		b
75.357	C	9	.575 s	Fe	3	218	218		V b4
76.788	B	17	7.007 s	Fe	4	219	218	+9	V
78.564	C	3	.768	Ti	1	204	218		II
83.689	B	18	.908	Fe	5	219	218	+5	V
84.826	B	19	5.040 s	Fe	6	214	218	+9	IV
87.067	A	18	.290 s	Fe	5	223	218	−3	V
91.382	C	9	.600		2	218	218		
5997.786	C	11	8.002	Fe	2	216	217		
6003.023	B	19	.239 s	Fe	6	216	217	+10	V
05.552	D	6	.770	Fe	1	218	217		
07.321	D	7	.540	Fe	1	219	217		
07.965	A	22	8.186 s	Fe	4	221	217	+6	
08.565	A	21	.785 s	Fe	6	220	217	+7	V
13.497	A	21	.715 s	Mn	6	218	216	+7	III
16.646	B	21	.861 s	Mn	6	215	216	+7	III
21.800	A	22	2.016 s	Mn	6	216	216	+3	
24.069	A	24	.281 s	Fe	7	212	212	+9	V
27.058	A	26	.274 s	Fe	4	216	215	+10	V b4
42.104	B	24	.315 s	Fe	3	211	215	+13	V e
56.005	B	27	.227 s	Fe	5	222	214	+1	V
6065.495	A	28	.709 s	Fe	7	214	214	+12	III b4

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
6078.499	A	26	.710 s	Fe	5	211	213	+3	V
79.016	A	23	.227 s	Fe	2	211	213		
82.717	C	17	.930	Fe	1	213	213		
84.121	D	3	.325		0	204	213		
85.255	C	20	.470	Fe-Ti	2	215	213		
86.287	B	17	.500	Ni	1	213	212		V
89.572	C	18	.787	Fe	1	215	212		
90.215	C	12	.429	Fe	2	214	212		
93.649	B	12	.864	Fe	2	215	212		
6096.673	B	17	.880	Fe	3	207	212		
6102.182	A	26	.392 s	Fe	6	210	212	+7	V
02.726	A	22	.937 s	Ca	9	211	212	+6	II
03.207	C	12	.400 s	Fe	4	193	212	+13	V
08.120	A	26	.334 s	Ni	6	214	211	−1	II
11.078	C	15	.290 s	Ni	2	212	211		
16.197	B	17	.397 s	Ni	4	200	211		V
22.226	B	21	.434 s	Ca	10	208	211	+10	II
25.021	C	8	.236		1	215	211		
26.228	B	8	.435	Ti	1	207	210	+11	II
27.912	B	22	8.124	Fe	3	212	210		b
28.983	C	9	9.190	Ni	1	207	210		
36.621	B	21	.829 s	Fe	8	208	210	+4	III b4
37.000	A	24	7.210	Fe	3	210	210		b
37.701	A	21	.915	Fe	7	214	210	+7	III b4
41.729	A	24	.938 s	Fe-Ba	7	209	210		V
45.021	B	14	.228		2	207	210		
49.248	B	17	.458		2	210	210		
51.623	A	24	.834	Fe	4	211	210		
54.229	A	12	.438 s	Na	2	209	210		
55.142	B	16	.350		7	208	210		
57.733	B	25	.945	Fe	5	212	210	+1	V b4
60.750	A	23	.956 s	Na	3	206	210		
61.294	B	21	.503	Ca	4	209	210		III
62.181	C	16	.390 s	Ca	15	209	210	+9	II
63.759	C	4	.968	Ca	3	209	210		III
65.363	B	24	.577	Fe	3	214	210		b
66.440	A	25	.651	Ca	5	211	211		III
69.040	A	25	.249 s	Ca	6	209	211		III
69.563	A	23	.778 s	Ca	7	215	211	+9	III
70.517	B	21	.730	Fe-Ni	6	213	211	+7	V
73.339	A	27	.553 s	Fe	5	214	211		III b4
6175.370	A	23	.584	Ni	3	214	211		V
76.816	A	27	7.027 s	Ni	5	211	212	+3	
80.208	A	26	.420 s	Fe	5	212	212		
86.717	C	13	.928	Ni	2	211	212		V
87.994	A	18	8.210	Fe	4	216	212		
91.181	A	26	.393 s	Ni	6	212	211		I
6191.570	A	20	.779 s	Fe	9	209	211	+11	II b4
6200.320	A	25	.527 s	Fe	6	207	210		IV b4
04.619	D	3	.825	Ni	1	206	210		III
13.436	B	26	.644 s	Fe	6	208	209	+3	III b4
15.147	A	16	.360	Fe	5	213	209		b4
16.359	B	12	.567	V?	1	208	209		
19.286	A	28	.494 s	Fe	6	208	209	±0	III b4
23.986	C	10	4.198	Ni	1	212	209		V
26.740	B	12	.951	Fe	1	211	209		
29.229	B	19	.437	Fe	1	208	210		
30.736	B	25	.943 s	V-Fe	8	207	210		III b4
32.645	A	27	.856	Fe	3	211	210	+1	V
37.322	C	16	.534 s		3	212	210		
38.387	B	14	.598		2	211	210		
40.652	B	20	.863	Fe	3	211	210		
43.114	D	6	.320	V	1	216	210		
43.813	C	12	4.033		2	213	210		
6244.476	C	12	.686		2	210	210		

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
6245.621	A	11	.832		1	211	210		
46.328	B	22	.535 s	Fe	8	207	210	+10	V
47.558	B	20	.774	V	2	216	210		
52.562	A	22	.773 s	Fe	7	211	210	+6	III b4
54.254	B	20	.456 s	Fe	5	202	210		III b4
56.366	A	22	.572 s	Ni-Fe	6	206	210		III b
58.110	B	20	.322	Ti	2	212	210	+4	II
58.712	B	19	.927	Ti	3	215	210	+6	II
61.103	A	19	.316 s	Ti	1	213	210	+2	II b4
65.138	A	19	.348 s	Fe	5	210	209	+8	
70.231	A	20	.442	Fe	3	211	209		III b
71.280	C	7	.486	Fe	0	206	209		
78.093	B	23	.303	Air	4	210	208		
78.881	A	8	9.084	Air	2	203	208		
79.102	A	9	.308	Air	3	206	208		
79.897	A	2	0.108	Air	2	211	208		
80.393	A	9	.598	Air	2	205	208		
80.623	C	6	.833	Fe	3	210	208		IA a
81.178	A	14	.387	Air	1	209	208		
81.955	A	15	2.164	Air	2	209	208		
82.730	A	3	.933	Air	2	203	208		
83.797	A	8	4.002	Air	1	205	208		
87.750	A	3	.953	Air	1	203	208		
89.398	C	10	.606	Air	1	208	208		
90.220	A	26	.427	Air	2	207	208		
90.967	C	13	1.184	Fe	4	217	208		e?
92.161	A	26	.373	Air	2	212	208		
92.959	A	26	3.170	Air	3	211	208		
95.178	A	23	.389	Air	3	211	209		
95.959	A	26	6.170	Air	3	211	209		
97.797	A	19	8.007	Fe	5	210	209	+7	III b4
6298.453	A	26	.666	Air	2	213	210		
6299.227	A	26	.436	Air	3	209	210		
6301.505	A	18	.718	Fe	7	213	211	+4	IV
01.998	A	21	.209	Air	2	211	211		
02.497	B	14	.709	Fe	5	212	211	+4	V
02.764	A	16	.975	Air	2	211	211		
05.810	A	20	6.024	Air	2	214	212		
06.564	A	20	.780	Air	2	216	212		
09.887	A	18	0.101	Air	2	214	213		
10.663	A	12	.848	Air	1	215	213		
14.663	B	16	.876	Ni	4	213	213	−3	II
15.314	B	13	.517	Fe	2	203	213		b
15.814	C	8	6.028	Fe	1	214	213		
18.025	B	17	.239	Fe	6	214	214	+7	III b4
18.710	A	4	.919		1	209	214		
22.693	B	19	.907	Fe	4	214	214	+5	III b
27.604	C	13	.820	Ni	2	216	215		II
30.098	B	12	.316	Cr	1	218	215		IA
30.852	C	11	1.067	Fe	2	215	215		
35.337	A	20	.554	Fe	6	217	215	+12	III b4
36.830	A	20	7.048	Fe	7	218	215	+6	V
38.874	B	5	9.096	Fe	2	212	215		
39.115	D	5	.335	Fe	2	220	215		
44.157	A	19	.371	Fe	4	214	215		III b
47.087	D	10	.310		2	223	215		
55.036	B	20	.246	Fe	4	210	215	+5	III b
58.685	B	18	.898	Fe	6	213	214		IA
62.869	B	7	3.090	Cr-Fe	2	221	214		
78.256	C	11	.468	Ni	2	212	211		IV
80.751	C	19	.958	Fe	4	207	211	+5	V b
6393.611	B	19	.820 s	Fe	7	209	211	+10	III b4
6400.005	B	17	.217 s	Fe	8	212	211	+3	III
00.326	C	13	.538 s	Fe	2	212	211		
6408.026	B	20	.233 s	Fe	5	207	211	+8	V

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
6411.656	B	20	.865 s	Fe	7	209	211	+8	IV
14.995	C	9	5.199		1	204	211		
16.930	C	11	7.133	Fe	1	203	211		
19.954	A	20	0.169 s	Fe	4	215	211	+5	V
21.360	B	21	.570 s	Fe	7	210	211	+10	III b
30.854	B	21	1.066 s	Fe	5	212	211	+10	III b4
32.686	B	13	.895	Fe	1	209	211		
39.085	A	22	.293 s	Ca	8	208	211	+13	II
49.820	B	23	0.033 s	Ca	6	213	212	+10	II
55.605	B	23	.820	Ca	2	215	213	+7	II
56.391	B	22	.603	Fe	3	212	213		
62.559	D	3	.784 s	Ca	5	225	214	−6	II
62.733	D	3	.965	Fe	3	232	214	+8	b
69.189	B	15	.408	Fe	2	219	216		V
71.668	A	23	.885 s	Ca	5	217	216	+8	II
75.632	D	7	.846	Fe	2	214	217		IV b
80.057	A	22	.285 s	Air	1	218	218		
81.875	B	25	2.098	Fe	3	223	218		IV
82.810	B	13	3.027	Ni	1	217	219		V
83.239	A	8	.468	Air	1	229	219		
86.777	C	3	7.005	Air	0	228	219		
6490.790	A	29	1.015	Air	1	225	220		
91.574	C	13	.800	Air	1	226	220		
92.903	A	13	3.130	Air	1	227	221		
93.789	A	31	4.004 s	Ca	6	215	221	+9	II
94.995	B	31	.213 s	Fe	8	218	221	+14	II b4
96.469	A	22	.688	Fe	2	219	222	+5	
96.911	A	33	7.128	Fe	4	217	222		
98.945	A	28	9.168	Fe	1	223	222		
6499.654	A	32	.880 s	Ca	4	226	222	+6	II
6504.193	D	3	.415	Air	0	222	223		
08.596	D	5	.826	Air	0	230	224		
14.730	A	28	.956	Air	2	226	226		
16.084	A	20	.311 s	Fe	2	227	226		
18.374	B	20	.599 s	Fe	2	225	227		b
23.839	B	12	4.080	Air	1	241	229		
32.352	C	13	.595 s	Air	1	243	232		
33.946	C	15	4.172 s	Air	2	226	232		
43.904	A	26	4.140	Air	2	236	235		
46.247	A	28	.479 s	Ti-Fe	6	232	235	+8	III b
47.692	B	10	.945	Air	0	253	235		
48.619	B	15	.855	Air	1	236	236		
52.626	B	20	.865 s	Air	1	239	236		
55.463	C	8	.700		1	237	236		
64.197	C	5	.450	Air	0	253	237		
69.223	A	28	.460 s	Fe	5	237	238	+9	V
72.083	C	11	.330	Air	1	247	238		
72.786	B	15	3.030	Ca	1	244	238		IA
74.234	C	13	.468 s	Fe	1	234	238		
75.022	C	12	.270 s	Fe	2	248	238		b
86.317	C	13	.550	Ni	1	233	239		II
92.926	A	29	3.161 s	Fe	6	235	240	+13	III b
93.881	A	29	4.121 s	Fe	4	240	240	+6	
6597.561	B	19	.807	Cr? Fe?	1	246	240		
6604.594	B	17	.837	Fe	1	243	240		
09.118	A	27	.360 s	Fe	3	242	241		IV b
27.551	C	12	.797	Fe	0	246	242		
33.753	A	27	.995 s	Fe	2	242	243	+1	V
43.638	A	31	.876 s	Ni	5	238	243	−3	I
63.451	A	30	.701 s	Fe	3	250	245		IV b
6677.996	B	31	8.235 s	Fe	5	239	246	+9	III b4
6703.569	C	17	.820 s	Fe	1	251	248		
05.108	C	19	.352 s	Fe	1	244	248		
15.387	D	4	.635	Cr? Fe?	1	248	248		
6717.686	B	20	.940 s	Ca	5	254	249		III

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
6721.849	C	10	2.096 s		2	247	249		
26.667	C	14	.925 s	Fe	2	258	249		
33.157	C	9	.410	Fe	1	253	249		
50.161	B	21	.407 s	Fe	3	246	250	+10	IV b
52.708	C	11	.965 s	Fe	1	257	250		
67.781	B	23	8.028 s	Ni	4	247	250	+3	I
6772.325	C	17	.568 s	Ni	2	243	251		V
6806.849	C	10	7.103 s	Fe	1	254	252		
10.268	B	21	.519 s	Fe	3	251	252		
20.375	B	13	.630 s	Fe	2	255	252		
28.600	B	21	.850 s	Fe	2	250	253		V
6839.826	C	8	.086	Fe	1	250	253		
41.344	A	23	.598 s	Fe	3	254	253		V
42.690	C	9	.945	Fe	1	255	253		
43.658	B	21	.913 s	Fe	3	255	253		V
55.169	B	24	.419 s	Fe	3	250	254	+10	V
57.250	D	4	.515	Fe	0	265	254		
58.154	A	24	.415	Fe	2	261	254		
62.496	D	4	.760	Fe	1	264	254		
67.221	A	26	.457 s	Air	6	236	254		
67.545	A	29	.800 s	Air	5	255	254		
69.936	C	4	.116	Air	7	180	254		
70.945	A	17	1.180 s	Air	8	235	242		
71.289	A	18	.532 s	Air	10	243	242		
72.247	A	31	.486 s	Air	11	239	242		
72.843	A	31	3.080 s	Air	12	237	243		
73.795	A	31	4.037 s	Air	12	242	243		
74.652	A	31	.899 s	Air	13	247	243		
75.592	A	31	.830 s	Air	13	248	243		
76.718	A	31	.958 s	Air	13	240	243		
77.637	A	30	.882 s	Air	12	245	243		
79.044	A	31	.288 s	Air	12	244	243		
79.929	A	34	0.172 s	Air	6	243	243		
83.833	A	31	4.076 s	Air	10	243	244		
85.756	A	31	6.000 s	Air	11	244	244		
86.744	B	31	.990 s	Air	12	246	244		
88.948	A	31	9.192 s	Air	13	244	245		
89.903	A	31	0.151 s	Air	14	248	245		
92.372	A	31	.618 s	Air	14	246	245		
93.310	A	31	.560 s	Air	15	250	245		
96.040	A	30	.289 s	Air	14	249	246		
96.965	A	31	7.208 s	Air	15	243	246		
6899.955	A	31	0.199 s	Air	14	244	246		
6900.869	A	31	1.117 s	Air	15	248	246		
04.118	A	31	.362 s	Air	14	244	246		
05.025	A	31	.271 s	Air	14	246	246		
08.535	A	30	.783 s	Air	13	248	247		
09.432	A	31	.676 s	Air	13	244	247		
13.201	A	33	.448 s	Air	11	247	247		
14.092	A	32	.337 s	Air	11	245	247		
14.573	B	20	.823 s	Ni	4	250	247		
16.686	B	21	.948 s	Fe	2	262	248		
18.121	A	33	.370 s	Air	9	249	248		
19.002	A	33	9.250 s	Air	9	248	248		
23.302	A	31	.553 s	Air	9	251	249		
24.173	A	31	.427 s	Air	9	254	249		
28.729	A	21	.977 s	Air	4	248	249		
29.595	B	13	.840 s	Air	4	245	249		
33.615	C	5	.887	Fe	2	272	250		
33.804	D	7	4.075	Air	2	271	250		
34.422	B	13	.670 s	Air	2	248	250		
35.280	B	14	.530 s	Air	2	250	250		
37.703	B	19	.957	Air	2	254	251		
39.618	B	13	.875	Air	2	257	251		
6940.190	B	6	.436	Air	2	246	251		

TABLE 1.—*A. O. B. S. solar wave lengths*—Continued

A. O. B. S. λ	Probable error	No. obs.	Rowland			Correction to Rowland		Sun—arc	Temperature and pressure classes
			λ	Iden.	Int.	Obs.	Mean		
6942.152	A	5	.402	Air	2	250	251		
43.802	B	15	4.060	Air	3	258	251		
45.214	B	17	.477	Fe	5	263	252	+10	IV
47.534	B	14	.782 s	Air-Fe	5	248	252		
51.248	B	13	.518	Fe	1	270	252		
56.426	C	7	.660 s	Air	4	234	253		
59.452	A	16	.704 s	Air	3	252	253		
61.260	A	16	.515 s	Air	4	255	254		
78.403	C	3	.670 s	Cr	1	267	257		
78.862	B	13	9.120	Fe	2	258	257	+4	IV
86.580	C	14	.833 s	Air	3	253	259		
88.518	D	4	.805	Fe	0	287	259		
88.986	B	14	9.237 s	Air	3	251	259		
90.373	C	7	.632	Air	1	259	259		
93.517	C	10	.776	Air	2	259	260		
94.107	D	7	.360	Air	2	253	260		
98.963	C	12	9.223	Air	2	260	261		
6999.882	C	9	0.155 s	Fe	1	273	261		
7003.574	C	4	.837	Si?	1	263	262		
04.747	C	12	.995	Air	2	248	262		
11.330	D	11	.590 s	Air	2	260	263		
22.956	B	8	3.230 s	Fe	2	274	265		
23.504	C	10	.770 s	Air	2	266	266		
27.479	B	10	.740 s	Air	2	261	266		
34.909	C	8	5.170 s		2	261	267		
38.219	C	9	.500 s	Fe	1	281	267		
39.796	B	11	.053 s	Air	1	267	268		
50.858	D	6	1.115	Air	0	257	269		
56.999	C	4	7.260	Air	0	261	269		
7084.981	C	4	5.265	Co	2	284	271		
7122.206	D	5	.484	Ni	4	278	275		IV
30.928	D	5	1.204	Fe	3	276	276	±0	
32.988	D	5	3.263	Fe	1	275	276		
7148.159	D	5	.435	Ca	3	276	277	+14	II

TABLE 2.—*Corrections to Rowland's wave lengths (3590–7142)*

Wave length	Correction	Wave length	Correction	Wave length	Correction
3590–3650	−0.141A	4165–4176	.163	4567–4587	.172
3651–3940	.140	4177–4187	.162	4588–4597	.173
3941–3952	.141	4188–4199	.161	4598–4634	.174
3953–3960	.142	4200–4213	.160	4635–4648	.173
3961–3970	.143	4214–4244	.159	4649–4671	.172
3971–3975	.144	4245–4295	.158	4672–4690	.173
3976–3978	.145	4296–4310	.159	4691–4708	.174
3979–3985	.146	4311–4320	.160	4709–4721	.175
3986–3992	.147	4321–4330	.161	4722–4729	.176
3993–3997	.148	4331–4357	.162	4730–4734	.177
3998–4000	.149	4358–4373	.161	4735–4738	.178
4001–4007	.150	4374–4378	.160	4739–4744	.179
4008–4018	.151	4379–4397	.159	4745–4749	.180
4019–4024	.152	4398–4414	.160	4750–4753	.181
4025–4028	.153	4415–4426	.161	4754–4760	.182
4029–4037	.154	4427–4435	.162	4761–4773	.183
4038–4090	.155	4436–4488	.163	4774–4785	.184
4091–4116	.156	4489–4500	.164	4786–4815	.185
4117–4123	.155	4501–4514	.165	4816–4825	.184
4124–4129	.154	4515–4523	.166	4826–4835	.183
4130–4135	.153	4524–4535	.167		
4136–4141	.152	4536–4550	.168	4836–4842	.182
4142	(1)	4551–4558	.169	4843–4848	.181
4143–4152	.163	4559–4562	.170	4849–4853	.180
4153–4164	.164	4563–4566	.171	4854–4860	.179

[1] Discontinuity.

TABLE 2.—*Corrections to Rowland's wave lengths (3590–7142)*—Continued

Wave length	Correction	Wave length	Correction	Wave length	Correction
4861–4865	−0.178	5497–5502	.206	6513–6516	.226
4866–4877	.177	5503–5506	.207	6517–6519	.227
4878–4884	.176	5507–5511	.208	6520–6522	.228
4885–4892	.175	5512–5515	.209	6523–6524	.229
4893–4900	.174	5516–5519	.210	6525–6528	.230
4901–4907	.173	5520–5524	.211	6529–6531	.231
4908–4920	.172	5525–5528	.212	6532–6534	.232
4921–5050	.171	5529–5532	.213	6535–6537	.233
5051–5067	.170	5533–5537	.214	6538–6541	.234
5068–5080	.169	5538–5543	.215	6542–6547	.235
5081–5095	.168	5544–5565	.216	6548–6553	.236
5096–5110	.167	5566–5604	.217	6554–6566	.237
5111–5130	.166	5605–5650	.218	6567–6581	.238
5131–5160	.165	5651–5660	.217	6582–6592	.239
5161–5175	.164	5661–5681	.216	6593–6604	.240
5176–5188	.163	5682–5700	.215	6605–6617	.241
5189–5200	.162	5701–5725	.214	6618–6631	.242
5201–5212	.161	5726–5737	.213	6632–6643	.243
5218	(1)	5738–5748	.212	6644–6656	.244
5224–5225	.155	5749–5762	.211	6657–6669	.245
5226–5230	.156	5763–5772	.210	6670–6681	.246
5231–5234	.157	5773–5795	.209	6682–6695	.247
5235–5238	.158	5805–5819	.214	6696–6718	.248
5239–5242	.159	5820–5870	.213	6719–6738	.249
5243–5246	.160	5871–5882	.214	6739–6767	.250
5247–5250	.161	5883–5894	.215	6768–6794	.251
5251–5254	.162	5895–5937	.216	6795–6822	.252
5255–5258	.163	5938–5972	.217	6823–6847	.253
5259–5262	.164	5973–5991	.218	6848–6869	.254
5263–5266	.165	5992–6007	.217	6870	(1)
5267–5271	.166	6008–6024	.216	6870–6872	.242
5272–5275	.167	6025–6044	.215	6873–6879	.243
5276–5280	.168	6045–6066	.214	6880–6886	.244
5281–5286	.169	6067–6084	.213	6887–6895	.245
5287–5291	.170	6085–6104	.212	6896–6907	.246
5292–5297	.171	6105–6123	.211	6908–6913	.247
5298–5302	.172	6124–6162	.210	6914–6920	.248
5303–5307	.173	6163–6176	.211	6921–6929	.249
5308–5311	.174	6177–6189	.212	6930–6937	.250
5312–5316	.175	6190–6196	.211	6938–6943	.251
5317–5320	.176	6197–6203	.210	6944–6952	.252
5321–5324	.177	6204–6240	.209	6953–6959	.253
5325–5329	.178	6241–6263	.210	6960–6964	.254
5330–5333	.179	6264–6272	.209	6965–6969	.255
5334–5338	.180	6273–6292	.208	6970–6974	.256
5339–5342	.181	6293–6297	.209	6975–6980	.257
5343–5346	.182	6298–6300	.210	6981–6985	.258
5347–5350	.183	6301–6303	.211	6986–6990	.259
5351–5354	.184	6304–6307	.212	6991–6996	.260
5355–5358	.185	6308–6315	.213	6997–7001	.261
5359–5361	.186	6316–6327	.214	7002–7006	.262
5362–5365	.187	6328–6355	.215	7007–7010	.263
5366–5369	.188	6356–6361	.214	7011–7016	.264
5370–5373	.189	6362–6368	.213	7017–7022	.265
5374–5376	.190	6369–6375	.212	7023–7028	.266
5377–5379	.191	6376–6444	.211	7029–7036	.267
5380–5383	.192	6445–6453	.212	7037–7049	.268
5384–5386	.193	6454–6459	.213	7050–7061	.269
5387–5389	.194	6460–6463	.214	7062–7073	.270
5390–5393	.195	6464–6467	.215	7074–7085	.271
5394–5396	.196	6468–6473	.216	7086–7096	.272
5397–5399	.197	6474–6477	.217	7097–7107	.273
5400–5402	.198	6478–6482	.218	7108–7118	.274
5403–5405	.199	6483–6486	.219	7119–7129	.275
5406–5408	.200	6487–6490	.220	7130–7141	.276
5409–5412	.201	6491–6495	.221	7142	.277
5413–5422	.202	6496–6499	.222		
5423–5438	.203	6500–6504	.223		
5439–5487	.204	6505–6508	.224		
5488–5496	.205	6509–6512	.225		

[1] Discontinuity.

WASHINGTON, May 23, 1928.

WAVE LENGTHS AND ZEEMAN EFFECTS IN YTTRIUM SPECTRA

By William F. Meggers

ABSTRACT

The wave lengths corresponding to approximately 1,000 lines photographed in the arc and spark spectra of yttrium were measured relative to secondary standards in the iron spectrum. The values extend from 2127.99A in the ultra-violet to 9494.81A in the infra-red. Comparison of relative intensities and other characteristics of lines in the different sources permitted a sharp discrimination between four classes of lines; about 500 are ascribed to neutral atoms (constituting the YtI spectrum), 240 originate with singly ionized atoms (YtII spectrum), 10 belong to doubly ionized atoms (YtIII spectrum), and most of the remainder describe the band spectrum characteristic of molecular compounds, presumably yttrium oxide. Measurements of Zeeman effects (photographed by Moore) for 220 yttrium lines ranging in wave length from 3173A to 6896A are included.

CONTENTS

	Page
I. Introduction	319
II. Experimental	320
1. Wave lengths	320
2. Zeeman effects	321
III. Results	321

I. INTRODUCTION

Wave-length measurements in the arc spectrum of scandium (Z=21) were published last year.[1] Many lines due to ionized atoms were observed in the arc, and on the basis of these measurements a fairly complete analysis of the ScI and ScII spectra was made.[2] Yttrium (Z=39) is a chemical analogue of scandium and should exhibit similar spectral characteristics, but attempts to analyze the structures of its spectra were only partially successful, and it was deemed essential to obtain new observational data. The observations reported in this paper consist of wave-length measurements and line intensity estimates in the ordinary arc and spark spectra of yttrium and additional data on the magnetic resolutions of the stronger lines.

[1] Meggers, B. S. Sci. Paper No. 549, 22, p. 61; 1927.
[2] Russell and Meggers, B. S. Sci. Paper No. 558, 22, p. 330; 1927.

II. EXPERIMENTAL

1. WAVE LENGTHS

Yttrium (Yt=89.33; Z=39), like scandium, is usually associated with some of the "rare earth" family of elements and is extremely difficult to obtain absolutely free from traces of the latter. This may account for a considerable number of the differences, especially among fainter lines, noted in various descriptions of yttrium spectra. The material used in the present investigation was purified with great care and patience by J. F. T. Berliner, of the Bureau of Chemistry, Department of Agriculture. Several grams of very pure yttrium oxalate kindly supplied for spectroscopic studies sufficed for numerous exposures for descriptions of arc and spark spectra and for the long exposures required in extending the observations of Zeeman effects. A portion of the same material was used by Doctor King and Miss Carter [3] for their study of the electric furnace spectra of yttrium.

The spectrograms from which the new wave-length values are derived were obtained by the same procedure which has been successfully used in other cases [4] where only small amounts of salts are available. Rods of pure silver were used as electrodes, yttrium salt was placed upon the lower electrode for the production of the arc spectrum, and the same rods were then used for the spark exposures. Sufficient yttrium was thus fused on the ends of the rods in the first case to give excellent spectra also in the second. Comparison arc and spark spectra of pure silver were photographed adjacent to those of yttrium and silver, so that lines due to the electrodes or to the atmospheric gases could be recognized at once, thus avoiding the necessity of measuring and subsequently eliminating these lines.

For the excitation of the arc spectrum an electric arc of 4 to 6 amperes direct current from a circuit with 220 volts potential difference was used. The spark spectra were produced by a 40,000-volt transformer consuming about a kilowatt, condensers of 0.006 microfarads capacity being placed in the secondary circuit in parallel with the spark.

The wave-length interval from 2500A in the ultra-violet into the infra-red (9500A) was investigated with large diffraction gratings. while the shorter wave portion to 2100A was photographed with a large quartz spectrograph. The concave gratings were mounted in parallel light as described in earlier publications.[5] For the red and infra-red regions a 6-inch grating rules by Anderson with 7,500 lines per inch, giving a scale of 10A per millimeter was employed, while the remainder of the range was recorded for the most part with a similar

[3] King and Carter, Astrophys. J., **65**, p. 86; 1927.
[4] Meggers, B. S. Sci. Paper, No. 499, **20**, p. 19; 1925.
[5] Meggers and Burns, B. S. Sci. Paper, No. 441, **18**, p. 191; 1922.

grating ruled by Rowland, with 20,000 lines per inch, which gave a dispersion of about 3.6A per millimeter in the first-order spectrum. The stronger lines between 3000 and 6000A were measured in the second-order spectrum of the latter grating, but many of the fainter lines, especially the somewhat hazy spark lines between 2500 and 3800A appeared only on spectrograms made with the first grating. The quartz spectrograph is one of the autocollimating type (E_1) made by Adam Hilger; its scale ranges from about 1.5A to 2.7A per millimeter in the interval of wave lengths for which it was used. The arc spectrum of iron was recorded with each yttrium spectrogram to supply the standard wave lengths from which the values for yttrium lines were derived by interpolation.

All the spectrograms were made on photographic plates of thin glass which could be bent to fit the focal curves of the spectrographs. The plates were sensitized [6] with pinaverdol, pinacyanol, dicyanin, or neocyanin to photograph the longer wave portions. Whereas previous attempts to photograph the arc spectrum of yttrium in the infra-red failed to record any lines of wave length exceeding 7881A, the use of neocyanin in the present case extended the wave-length data to 9495A. The exposure times ranged from a few minutes for the ultra-violet to an hour for the intra-red.

2. ZEEMAN EFFECTS

The spectrograms for the study of Zeeman-effects of yttrium lines were all made in 1924 at the Brace Laboratory of Physics, University of Nebraska, by the late Prof. B. E. Moore. The exposures were made on long strips of Eastman film adjusted to the focal curve of a concave grating spectrograph giving a scale of about 2.5A per millimeter in the first order. A 5,000-volt transformer was employed by Moore for producing sparks between carbon plates impregnated with yttrium solutions and inserted between the pole pieces of the electromagnet. The magnetic field strength was of the order of 28,000 gausses per square centimeter; it was determined from the separations of the magnetic components of sodium lines (D_1 and D_2) or of calcium lines (H and K) which were present as impurities. Separate exposures were made for the parallel and for the perpendicular components; all of the measurements and calculations were made at this bureau.

III. RESULTS

Various observers have already described limited portions of the arc and spark spectra of yttrium. The most reliable results up to the year 1911 are quoted by Kayser in his Handbuch der Spectroscopic, Volume VI; they are by Kayser [7] (arc spectrum 2227.849 to

[6] Walters and Davis, B. S. Sci. Paper, No. 422, **17**, p. 353; 1921.
[7] Kayser, Abhandl. Berlin Akad., **36**, p. 633; 1903.

6701.188A), by Eberhard [8] (arc spectrum 2760.17 to 4527.95A), by Eder and Valenta [9] (arc spectrum 5466.669 to 6815.6A), and by Exner and Haschek [10] (arc spectrum 2422.30 to 6795.71A and spark spectrum 2191.35 to 6795.70A). These values are all based on Rowland's system of standard wave lengths.

More recent measurements have been made on the international scale of wave lengths by Eder [11] (arc spectrum 2231.55 to 7881.69A), by Kiess [12] (arc spectrum 5503.474 to 7881.868A), and by Yntema and Hopkins [13] (arc spectrum 2243.02 to 4199.26A). Admitting that these observations are superior in many respects to the earlier ones, it is, nevertheless, obvious that they have certain shortcomings as descriptions suitable for a complete analysis of arc and spark spectra. They all refer to the arc spectrum without any indication as to which lines belong to ionized atoms, and, furthermore, they cover their respective wave-length intervals with somewhat different scales of intensities and with many disagreements as to the fainter lines.

The only extensive data on Zeeman effects of yttrium lines are those published by Moore [14] in 1908. These give complex patterns for 12 lines; 14 lines were observed as quartets in the magnetic field and 74 as triplets. The lines range from 3130A to 5663A. A few additional observations were published by Meggers and Moore [15] in 1925.

The results of our determinations of approximately 1,000 wave lengths in the arc and spark spectra of yttrium and of Zeeman effects for 220 lines are presented in Table 1. Wave lengths on the international angstrom scale appear in the first column, and the estimated relative intensities of the lines in the arc and spark are given in the second. Certain other symbols appearing in the intensity column are explained below.

The probable errors of my wave-length determinations are usually less than 0.01A for the stronger lines between 3000 and 6000A, but the errors for the remaining lines are somewhat larger, since they were measured for the most part on smaller scale spectrograms. It was especially difficult to obtain accurate values for lines marked *h*, *l*, *hl*, or *nl;* because of their broad and unsymmetrical character the measured effective wave length depends somewhat on the exposure; that is, lines shaded toward the red were invariably measured as having longer wave lengths in stronger exposures. The probable error for such lines averages about 0.05A.

---

[8] Eberhard, Zeitschr. wiss. Phot., **7**, p. 245; 1909.
[9] Eder and Valenta, Kaiser, Akad. wiss. Wien. Berlin, **190**, p. 9 and p. 519; 1910.
[10] Exner and Haschek, Die Spectren der Elemente, Deuticke, Leipzig; 1911 and 1912.
[11] Eder, Kaiser, Akad. Wiss., Wien, Ber., **125**, p. 383; 1916.
[12] Kiess, B. S. Sci. Paper, No. 421, **17**, p. 318; 1921.
[13] Yntema and Hopkins, J. Opt. Soc. Am., **6**, p. 121; 1922.
[14] Moore, Astrophys. J., **28**, p. 1; 1908.
[15] Meggers and Moore, J. Wash. Acad. Sci., **15**, p. 207; 1925.

My intensity estimates were made on an expanded scale more or less like that developed by King. Experience in classifying the lines in complex spectra has shown that such large-scale intensities are helpful in detecting multiplet structures, while the 1 to 10 scale which most spectroscopists have used is not very instructive. Comparison of estimated intensities in arc and spark spectra, especially when these are photographed side by side as in the present case, enables one to decide if the line belongs to spectrum I of neutral atoms, to spectrum II of singly ionized atoms, or to spectrum III of doubly ionized atoms. Lines of the YtII spectrum may be divided roughly into two classes—those which are nearly as strong in the arc as in the spark and those which appear weak in the arc but greatly enhanced in the spark. The latter are usually hazy and unsymmetrical in the spark. Lines belonging to doubly ionized atoms, YtIII are distinguished by being very strong in the spark, but either absent or extremely weak in the arc.

A small number of lines observed by Exner and Haschek and confirmed by King are included in column 1.

For purposes of comparison and to illustrate certain points of interest Eder's wave lengths and intensities for lines observed in the arc spectrum are given in the next two columns. After correcting some obvious typographical errors in Eder's wave lengths (4039, 4077, 4124 instead of 4030, 4076, 4025) there is, in general, good agreement between his values and mine for the stronger lines, but there are some unaccountable omissions in his list; for example, 3045, 3776, 4487 and a considerable number of faint lines present in his but absent from mine. Similar discrepancies as to the faint lines are noted in comparing any list of yttrium lines with any other of those mentioned above. Some of them may arise from differences in judgment in picking lines out of superposed band structures and certain others may represent unidentified impurities. Special attention is called to the systematic wave-length differences for lines which are shown by my intensity estimates to be hazy, unsymmetrical enhanced lines. For these the effective wave length as measured in the spark is usually from 0.1 to 0.2A greater than the value obtained from arc spectra in which most of the lines appear also but with relatively low intensity. A similar displacement of these enhanced lines was noted much earlier by Exner and Haschek [16] in their first description of the arc and spark spectra.

In column 5 of Table 1 the arc intensity estimates and furnace temperature classes published for yttrium lines by King and Carter [17] are quoted. Lines in Classes I and II appear at low temperature, 2,000° C. Those of Class I show a slower change from low to high

[16] Exner and Haschek, Wellenlangentabellen, Deuticke, Leipzig; 1902.
[17] King and Carter, Astrophys. J., **65**, p. 86; 1927.

temperature than those of Class II and as a rule are less conspicuous in the arc. Lines of Class III are usually well developed at medium temperatures, 2,200 to 2,300°, while lines clearly associated with high temperatures, 2,600 to 2,800°, are placed in Classes IV and V, those of Class V being absent or very faint in furnace spectra. The other symbols in column 5 have significance as follows: *d*, unresolved doublet; *n*, diffuse arc line; *A*, relatively stronger in the furnace than in the arc; *E*, enhanced line.

The vacuum wave numbers corresponding to the observed yttrium wave lengths are presented in column 6 of Table 1. They were taken from Kayser's Tabelle der Schwingungszahlen, and usually represent the mean wave length from columns 1 and 3, but in a few cases where considerable divergence exists the value in column 1 has been preferred.

Zeeman effects for 220 yttrium lines are found in column 7. These represent the more recent observations. They are not only more extensive, but also somewhat better quality than those published earlier. The patterns are presented in the standard notation for Zeeman effects; that is, the separations are expressed in decimal parts of *a*, the separation of a normal triplet, components polarized parallel to the magnetic field being inclosed in parentheses and followed by the perpendicular components. In complex patterns the strongest component of the group is printed in boldface type. A few lines for which the focus and exposure were best were just barely resolved when the neighboring components were separated by about 1/5*a*, but many strong lines were regularly overexposed, so that the components could not be resolved even when separated by 1/4*a* or 1/3*a*. For unresolved patterns an effort was made to measure the center of gravity of the unresolved group and to give some indication of the intensity distribution among the fused components. For this purpose, the notation used by Back [18] for distinguishing various types of intensity gradients is employed here. The letters A or B after a Zeeman effect mean that the pattern is complex but unresolved, and A indicates that the maximum intensity for perpendicular components is at the edge of a group, while B signifies that it is in the middle of the group. The distinction between strongest component inside or outside of the group is shown by A^1 and A^2, respectively. The complete interpretation of these Zeeman effects will be given in another paper [19] dealing with the spectral-series classification of yttrium lines.

In the last column an attempt is made to assign each observed wave length to its proper atomic or molecular source. This separation of lines into YtI, YtII, YtIII spectra, and assignment of bands to molecular orgin, is based primarily upon the relative intensities and

[18] Back, Zeitschr. fur Phys., 15, p. 212; 1923.
[19] Meggers and Russell, forth coming paper in B. S. Journal of Research.

other characteristics reported in column 2. It is supported by the temperature classification and further description in column 5 and by the Zeeman effects describing combinations of spectral terms having even multiplicity for YtI lines and odd multiplicity for YtII lines. The detailed correlation of all the descriptive data on yttrium lines presented above is reserved for a subsequent paper on the analysis of the arc and spark spectra of yttrium. The meaning of the symbols and abbreviations used in Table 1 is summarized as follows:

h=hazy (=n in King's column).
l=shaded to long wave lengths.
n=B. H.=band head.
p=part of band structure.
d=double.
c=complex.
E=enhanced line.
A=stronger in furnace than in arc.
A^1=strongest s-components of Zeeman effect inside.
A^2=strongest s-components of Zeeman effect outside.
B=strongest s-components of Zeeman effect in center.
w=wide.
$E+H$=Exner and Haschek.

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
2127.99	—	100				46977.8		III
2191.22	—	200				45622.4		III
2200.80	—	50				423.8		III
06.22	—	30				45312.2		III
31.	—	—	.55	1		44797.9		
43.06	25	50	.03	2		568.4		II
68.14	—	2h				44075.8		II
2284.5	—	100hl				43759.7		III
2327.30	—	20				42955.0		III
28.	—	—	.95	1		924.6		
31.	—	—	.63	1		875.3		
32.	—	—	.58	2		857.8		
40.8	1	10h	.79	1		707.4		II
49.	—	—	.69	1		545.8		
54.	—	—	.20	3		464.3		
55.	—	—	.40	1		442.6		
58.	—	—	.70	2		383.3		
61.	—	—	.81	2		327.5		
67.25	3	200				42230.2		III
85.	—	—	.24	2		41911.7		
2398.14	1	10hl	.06	2		686.8		II
2404.			.11	1		582.8		
13.92	—	3h	.94	1		413.6		II
14.68	5	100				400.8		III
17.4	1	5h	.29	1		355.0		II
22.22	20	50	.20	4		41272.1		II
57.			.93	½		40672.3		
60.			.11	1		636.3		
60.62	2	20	.60	2		628.0		II
2465.90	—	5h				40540.9		II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
2479.			.09	1		40325.2		
79.			.80	1		313.7		
2490.4	1	—				40142.1		I
2529.			.14	1		39527.3		
40.			.28	1		353.9		
47.			.56	1		241.5		
50.			.35	1		198.5		
54.			.87	1		39129.2		
64.3	—	1				38985.3		II
70.72			.72	1		888.0		
79.			.36	1		757.7		
93.			.76	1		542.5		
2594.			.88	1		525.9		
2612.			.38	1		267.9		
19.			.46	1		38164.4		
34.			.32	1		37949.2		I
47.			.74	½		756.8		
71.			.20	½		425.2		
72.			.08	1		412.9		
81.			.65	1		279.4		I
84.			.20	½		244.0		I
94.			.21	1		105.6		
2695.			.40	1		37089.2		
2705.			.85	1		36946.0		
10.			.15	1		887.4		
19.			.99	1		754.0		
23.			.00	3		713.3		I
30.			.06	1		618.4		I
33.			.93	1		566.6		
34.98	—	4h	.85	2		553.3		II
42.			.55	3		451.7		I
49.4	—	1	.23	1		361.9		II
50.40	—	3h	.20	2		348.9		II
55.			.79	1		276.5		
56.			.33	1		269.4		
60.			.10	3		36219.9		I
85.23	—	3	.19	2		35893.4		II
85.60	—	2	.58	2		888.5		II
2791.			.20	1		816.3		
2800.11	—	4	.12	2		702.3		II
07.			.66	1	1 III A	606.4		I
13.61	—	4h	.66	1	8 III	531.1		I+II
17.01	5	200	—	—		488.2		III
18.			.87	1	3 IV	464.8		I
22.			.56	1	10 IV	418.4		I
23.			.55	1		406.0		
24.			.48	1		394.3		
25.37	—	3h	—	—		383.2		II
26.38	1	5	—	—	1 IV A	370.6		II
34.57	—	5h	.39	1		269.5		II
40.98	1	5h	.84	1		189.7		II
42.5	—	1	—	—		170.0		II
50.7	—	2	—	—		068.8		II
54.45	2	15	.42	2	3 V E	022.9		II
56.32	1	6	.30	2	1 V	35000.0		II
57.			.87	1		34980.8		
58.06	1	4h	—	—		978.5		II
71.4	—	1h	.20	1		817.2		II
78.92	—	1	—	—		725.1		
83.85	—	2	—	—		665.7		
86.			.49	2	15 III	634.0		I
90.			.46	1	3 IV	587.2		I
91.			.32	1		576.2		
[97.70	1	5	.68	1	1 V	500.2		II
2898.93	1	3	.82	1	tr V	34486.0		II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
2901.			.48	1	6 III	34455.1		I
07.18	—	2	—	—		387.5		II
19.05	3	—	.06	3	20 II	247.7		I
29.			.00	1		131.4		I
30.15	1	6h	.03	2		118.7		II
30.8	—	2h	.77	1		110.7		II
35.			.91	1		34051.1		
43.			.58	1		33962.3		
45.92	4	150	—	—		935.4		III
48.39	5	1	.40	4	30 II	906.9		I
48.98	—	3h	.78	1		901.3		II
50.33	—	1h	—	—		884.6		II
53.28	—	3h	.14	1		851.6		II
56.04	1	5h	5.86	1		820.2		II
57.39	—	2h	—	—		803.7		II
64.96	5	1	.95	3	30 II	717.5		I
74.02	1	5h	3.91	1		615.4		II
74.59	6	1	.60	4	35 II	608.2		I
78.18	—	3h	7.99	1		568.6		II
80.7	2	20hl	.55	2	2 V E	540.3		II
82.20	—	2	—	—		522.5		II
84.25	10	2	.25	4	50 II	499.5		I
95.26	1	—	.25	2	10 III	376.4		I
2996.94	2	—	.94	3	20 III	357.7		I
3001.42	—	2	—	—		307.9		II
05.26	1	—	.25	2	12 III	265.4		I
06.0	1	2h	—	—		257.1		II
09.			.51	1		218.4		
18.			.95	2	6? III	114.5		I
21.76	2	—	.73	3	15 II	083.9		I
22.30	2	—	.27	3	12 II	077.9		I
23.50	—	2h	—	—		064.7		II
23.			.70	1		062.5		
23.			.99	1		059.3		
26.5	2	10hl	—	—		031.9		II
27.75	—	3	.68	1		33018.6		II
30.2	1	4h	.08	1		32992.2		II
36.7	3	25hl	.59	3		921.5		II
37.			.82	1		908.8		
38.			.46	1	4 III A	901.9		I
39.			.98	1		885.4		
44.			.84	2	5 III	832.9		I
45.36	4	1	—	—	20 II	827.3		I
47.			.11	1	3 III	808.5		I
47.			.41	1	3 III	805.2		I
49.			.86	1		778.9		
50.5	—	1h	—	—		772.0		II
51.			.52	1		761.0		
53.3	2	15hl	.26	2		742.2		II
54.			.41	1		730.1		
55.3	4	50hl	.21	3		721.0		II
56.			.33	1	2? III ?	709.5		I
59.			.50	2Dy	4 III	675.6		I
66.02	—	4h	5.83	1		607.2		II
67.			.27	1		592.8		
69.26	—	5h	.04	1		572.9		II
72.			.32	2	5 III	539.3		I
76.			.49	2	5 III	495.2		I
77.14	—	4h	6.95	1		489.3		II
78.64	—	4h	.57	1	2 V E	472.9		II
81.6	—	2h				441.3		II
82.16	—	3h	.16	1Al		435.4		II
86.9	5	30hl	.84	4	7 V	385.9		II
91.			.70	3	15 III	335.3		I
3093.76	1	10h	.75	3		32313.8		II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
3095.			.49	1		32295.7		
95.88	3	4	.88	4	25 V E	291.6		II
3096.			.57	1	3 III	284.5		I
3103.3	—	2h	.25	1Dy?		214.7		II
03.			.09	2	4 IV	210.4		I
04.82	—	4h	.69	2		199.3		II
08.			.86	2	5 III	156.8		I
09.3	—	1	—	—		152.3		II
09.			.77	1Dy?		147.4		
10.65	—	2h	.50	1		139.0		II
11.80	1	—	.80	3	15 III	126.5		I
12.05	2	4	.03	3	18 V E	124.0		II
14.			.27	3	2 III	101.0		I
14.45	—	10h				099.1		II
18.			.50	1Ho?		057.4		
22.			.60	1		32015.3		
26.16	—	4h	.00	1	5 V E	31979.7		II
28.8	3	20hl	.74	3	5 V E	952.2		II
30.0	5	40hl	29.93	4	10 V E	940.0		II
33.			.15	1		907.6		
35.17	4	5	.16	4	25 V E	887.0		II
40.			.63	1Dy?		831.6		
41.			.16	1Ny?		826.2		
44.37	—	2h	.20	1		794.6		II
52.			.67	2	15n III	710.0		I
55.62	(1)	E+H	—	—	3 III A	680.4		I
57.			.50	1		661.5		
58.			.36	1		652.9		
59.			.47	1		641.7		
60.60	—	1h	.54	1Dy?		630.7		II
62.			.83	1Dy?		608.1		
64.			.76	1		588.9		
70.			.00	1Dy?		536.6		
71.			.69	2	4 III A	519.8		I
72.85	(1)	E+H	—	—	4 III A	508.3		I
73.07	6	100hl	.05	4	10 V E	506.2	(0.00)—	II
73.			.72	1		499.7		
74.			.36	1		493.3		
79.42	5	10	.40	4	40 V E	443.3	(0.00)—	II
82.42	1	3hl	.23	2		414.6		II
85.			.93	1	2 III A	379.0		I
88.			.75	1		351.2		
91.34	2	—	.29	3	15 III	326.0		I
93.48	—	2hl	.29	2		305.4		II
94.			.37	2	6 III	296.3		I
95.62	25	50	.61	6	100 III E	283.8	(0.95) 1.40	II
97.			.69	1		263.6		
3198.5	—	2h	.41	2		256.0		II
3200.28	25	50	.25	6	100 III E	238.4	(w) 1.26B	II
03.33	30	60	.32	6	100 III E	208.6	(0.00) 0.46	II
03.			.82	1		203.7		
06.			.22	1		180.4		
09.			.35	2	3 III A	150.0		I
11.			.26	1		131.5		
12.40	1	5hl	.28	2		121.0		II
14.			.04	1		104.5		
15.			.20	1Dy?		093.3		
16.70	50	100	.67	10	150 III E	079.0	(0.00) 0.98	II
17.			.80	1		068.2		
20.			.72	1		040.0		
21.			.50	1Dy?		032.5		
22.			.02	1		027.5		
23.			.28	1Dy?		31015.4		
25.17	1	5hl	.03	3		30997.9		II
3227.			.08	1		30978.9		

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
3227.			.69	1		30973.0		
31.20	—	2h	—	—		939.4		II
31.			.32	1		938.2		
32.00	—	3h	1.80	2		932.6		II
37.			.93	1		875.5		
39.			.29	1		862.1		
42.30	60	150	.28	15	200 III E	833.5	(0.00) 1.18	II
47.			.02	1		788.6		
51.			.29	2Dy?		748.2		I
52.28	1	—	.27	3	8 III	738.8		I
55.			.82	1		705.4		
56.			.20	1		701.8		
57.			.52	1		689.4		
61.			.23	1		654.5		
63.			.22	1		635.8		
64.			.77	3Ho?		621.2		
67.			.24	1		598.1		
67.			.81	1		592.7		
69.			.11	1		580.6		
69.			.40	1		577.9		
70.			.94	1		563.5		
71.			.13	1		561.7		
73.			.04	1		543.8		
75.			.56	2	2 IV	520.4		I
78.			.43	2	5 III A	493.6		I
79.			.35	1Er?		485.1		
80.			.13	2Dy?		477.8		
80.	?Ag	?Ag	.91	4	10 V E	470.6		II
81.			.98	1Ho?		460.7		
82.51	—	2	.45	3		456.0		II
82.			.77	1		453.3		
83.			.21	2		449.2		
83.			.85	1		443.3		
86.71	—	3h	.68	3		416.9		II
87.			.21	3	4 III	412.2		I
87.			.93	1Dy?		405.5		
90.			.11	1		385.4		
90.			.56	3		381.2		
90.			.96	1Ho?		377.5		
91.			.44	1Dy?		373.1		
93.			.44	2		354.7		
93.9	—	3h	.68	2		351.4		II
94.			.55	1		344.4		
3298.			.26	1		310.3		
3302.			.17	2		274.4		
02.			.56	½		270.8		
04.0	—	2h	3.86	1		258.3		II
04.			.32	½		254.7		
05.			.49	½		243.8		
05.			.90	½		240.3		
06.			.27	½		236.9		
07.			.61	½		224.6		
08.5	2	20hl	.47	3		216.7		II
08.			.84	1		213.4		
10.			.13	½		201.6		
12.5	—	4hl	.40	1		180.5		II
12.			.67	½		178.5		
15.			.40	½		153.6		
16.			.32	½		145.3		
17.			.03	½		138.8		
18.6	1	4hl	.52	2		124.9		II
19.8	1	15hl	.76	3		113.8		II
20.			.60	1		106.4		
23.			.18	1		083.0		
3327.89	50	100	.89	15	150 III E	30040.4	(0.00) 1.00	II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
3330.9	2	20hl	.88	2	3 V E	30013.4		II
33.6	—	2h	.42	1		29989.8		II
35.			.20	2	8n V	974.6		I
36.3	1	4hl	.18	1		965.3		II
37.85	1	—	.82	2		950.9		I
38.			.76	1		943.7		
40.			.37	3	7 III	928.2		I
40.			.98	½		922.8		
41.			.85	1		915.0		
44.			.53	2	4n V	891.0		I
49.			.26	1		848.8		I
52.			.64	½		818.7		
53.			.56	½		810.5		
54.			.57	2	} 10d? III {	801.5		I
54.81	(1)	E+H	—	—		799.4		I
58.			.94	2	5 III	762.8		I
62.00	4	30hl	1.99	5	20 V E	735.7	(0.00) 1.10	II
64.77	1	—	.79	2	7 III	711.1		I
77.			.72	2	6 III	597.3		I
80.1	—	5hl	—	—		576.5		II
82.			.83	½	5 V	552.6		I
83.			.06	1	7 V	550.6		I
88.59	1	—	.58	2	15 III	502.3		I
89.81	(2)	E+H	—	—	5 IV ?	491.7		I
94.97	(2)	E+H	—	—	4 III	446.9		I
3397.05	2	—	.03	3	15 III	429.0		I
3407.7	—	3h	—	—		336.9		II
07.82	2	—	—	—		335.9		
09.9	1	4hl	—	—		318.0		II
12.47	2	—	.47	2	12 III	295.9		I
14.5	(2)	E+H	—	—	8n III	278.5		I
29.4	—	3h	—	—		151.3		II
31.01	(2)	E+H	—	—	6 V	137.6		I
31.67	(2)	E+H	—	—	5 III	132.0		I
33.			.02	1		120.6		
33.79	(2)	E+H	—	—	6n IV	114.0		I
37.95	1	—	—	—	8n IV	29078.8		I
48.82	8	10	.81	4	35 V E	28987.2	(0.72) 1.20 Bw	II
50.95	2	—	.94	2	12 III	969.2		I
55.94	} 4c	1	{ —	—		927.4		
56.10			{ —	—		926.1		
57.1	1	4hl	—	—		917.7		II
61.0	2	20hl	—	—	4 V ?	885.1		II
67.88	4	5	.88	4	20 V E	827.8		II
70.3	—	5hl	—	—		807.7		II
71.7	1	—	—	—	10n V ?	796.1		I
73.18	(2)	E+H	—	—	5 III	783.8		I
84.06	2	—	.06	2	10 III	694.0		I
84.89	2	—	—	—		687.1		
85.73	4	1	.73	4	40 III	680.2	(0.00) 1.15	I
96.08	40	80	.09	8	150 III E	595.3	(0.00) 0.62	II
3498.94	2	—	—	—	10n III	571.9		I
3500.60	1	—	—	—	7 III	558.4		I
01.95	2	—	—	—	2 IV	547.4		I
06.47	(2)	E+H	—	—	6 III	510.6		I
08.0	2	8hl	—	—		498.2		II
10.54	1	—	—	—	8 III	477.5		I
11.19	2	—	.20	3	15 III	472.2		I
12.88	3	—	.90	2	20 III	458.5	(—) 1.50 A^3	I
21.54	1	—	—	—	6 III	388.6		I
31.71	3 Dy?		.65	2		307.1		
44.0	1	3hl	.03	4		208.5		II
44.			.93	3		201.3		
46.0	2	5hl	—	—		192.8		II
3549.02	50	100	8.99	6	150 III E	28168.9	(0.00) 1.52 A^2	II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
3551.80	2	—	—	—	8 III ?	28146.7		I
52.70	10	2	.69	4	40 I	139.6	(0.80) 0.45, 1.00, 1.54	I
56.1	1	5hl	—	—		112.7		II
58.76	2	1	—	—	5 II	091.7		I
59.65	1	1	—	—		28084.7		
71.43	4	2	.44	1	10 II	27992.0	(—) 1.22	I
73.77	2	2	—	—		973.7		
76.06	5	2	.04	2	12 III	955.8	(—) 1.30	I
84.53	60	100	.51	4	150 III ? E	889.8	(**0.00**, 0.60) 0.54, 1.15, **1.74**	II
87.75	5	2	.75	1	20 III	864.7	(0.00) 0.95	I
89.68	2	—	—	—	5? III ?	849.7		I
89.91	3	2	—	—		847.9		
90.30	2	—	—	—		844.9		
3592.93	25	5	.91	4	200 II	824.6	(0.00) 0.84	I
3600.74	100	300	.72	6	500 III E	764.2	(0.00) 1.31	II
01.93	75	100	.91	5	200 III E	755.1	(0.00) 0.55	II
05.4	2	10hl	—	—		728.3		II
08.			.84	1		701.9		
11.06	100	200	.05	10	400 III E	684.9	(0.00) 1.16	II
12.34	2	—	—	—		675.0		
12.70	2	—	—	—		672.3		
18.8	1	—	.77	1		625.8		I
20.95	50	10	.94	6	400 II	609.3	(0.00) 1.10 A^1	I
21.86	2	1	—	—		602.3		
22.19	2	1	—	—		599.8		
28.71	50	100	.70	5	150 III E	550.2	(**0.00**, 0.59) 0.58, 1.16, **1.72**	II
33.13	100	200	.11	8	300 III E	516.7	(0.00) 0.91	II
35.4	3	20hl	.32	1		499.8	(w ?) 1.03	II
39.28	1	—	.27	3		470.2	(—) 1.39	I
40.34	1	1	—	—		462.2		
43.4	1	3hl	—	—		439.1		II
45.40	2	1	—	—		424.0	(—) 1.32	
50.45	—	2h	—	—		386.1		II
53.60	2	2	—	—		362.5		II
64.62	100	150	.59	10	200 III E	280.3	(0.00) 1.50 A^2	II
65.75	1	—	—	—		271.8		I
68.5	5	50hl	.48	3	3 V E	251.4	(0.00) 1.07	II
75.6	1	5hl	—	—		198.7		II
84.9	1	5hl	—	—		130.1		II
89.2	—	2h	—	—		098.5		II
92.52	3	1	.54	6	8 III	074.0		I
94.80	1	—	—	—		057.4		
96.6	3	25hl	—	—		044.2	(—) 1.12	II
3699.14	3	3	—	—		27025.6		II
3702.84	2	1	—	—		26998.6		
03.3	1	5hl	—	—		995.3		II
10.30	200	500	.30	15	800 III E	944.4	(w) 1.15 A^1	II
14.3	1	5hl	—	—		915.3		II
17.0	1	7hl	6.94	1		896.0		II
18.09	2	—	.14	3	5 IV	887.7		I
21.40	3	4	—	—		864.0		II
24.			.76	2		839.8		
27.0	2	20hl	—	—		823.6		II
38.61	2	—	.62	2	4 III	740.3		I
47.55	30	40	.55	3	20 III E	676.6	(0.44) 0.48, 0.94	II
49.89	2	—	—	—		659.9		I
58.9	1	3hl	—	—		596.0		II
62.97	2	3	—	—		567.2		
74.33	150	300	.33	5	200 III E	487.3	(0.00) 1.05	II
3776.56	50	75	—	—	40 III E	26471.6	(0.00) 1.31	II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
3777.28	2	2	—	—		26466.6		
82.3	5	5hl	—	—		431.5		II
85.62	3	3	—	—		408.3		II
88.70	100	200	.69	5	100 III E	386.8	(0.00) 0.82	II
3792.5	2	10hl	—	—		360.4		II
3800.03	2	1	—	—		308.1		
00.9	2	15hl	—	—		302.1		II
08.7	—	1h	—	—		248.3		II
12.0	1	5hl	—	—		225.5		II
13.8	—	2hl	—	—		213.2		II
18.34	40	60	.37	3	20 III E	181.9	(0.36, **0.76**) 0.40, **0.79, 1.20,** 1.57	II
24.8	1	5hl	—	—		137.8		II
25.			.91	1		130.2		
32.89	50	100	.87	2	30 III E	082.7	(0.66) 0.69, 0.90, **1.10, 1.30,** 1.51, 1.72	II
40.			.43	1		26031.4		
46.5	1	3h	—	—		25990.3		II
48.2	?	8hl	—	—		978.8		II
72.3	?	5hl	—	—		817.2		II
78.28	15	20	.27	1	20 V ? E	777.4		II
87.76	2	—	—	—	5 III	714.5		I
87.			.93	2		713.4		
90.			.13	1		698.8		
90.86	4c	—	—	—		694.0		
91.08			—	—		692.6		
92.			.41	2		683.8		I
3896.8	1	10hl	—	—		654.8		II
3900.			.27	1		632.0		I
04.59	2	—	.59	2	5 III	603.7		I
18.24	2	—	—	—	4 III	514.5		I
30.10	3	—	—	—		437.5		I
30.66	15	15	.65	3	10 V E	433.8	(w) 1.06 B	II
38.47	2	1	—	—		383.4		
42.48	1	—	—	—		357.6		I
44.68	2Dy?	—	—	—		343.4		I
46.21	3	2	.20	2		333.7		II
50.35	150	200	.35	5	50 III E	307.1	(**0.00**, 0.46) 0.45, 0.97, **1.49**	II
51.59	10	5	.60	3	8 V E	299.1	(w) 1.28 A^1	II
55.08	3	—	.09	3	5 III	276.8		I
62.19	2	1	—	—		231.4		
67.7	3	15hl	—	—		196.4		II
68.42	5Dy?	10	—	—		191.8		
70.62	2	1	—	—		177.9		
73.	—	—	.45	2		160.0		
74.61	1	—	—	—		152.6		
78.59	2Dy?	2	—	—		127.4		
82.59	100	150	.60	8	100 III E	102.2	(0.28) 1.03 B	II
87.48	2	—	.50	1	4 III	071.4	(—) 1.00	I
94.52	2	—	—	—		027.2		
3997.43	1	1	—	—		25009.0		II
4029.84	2	—	.86	1	3 IV	24807.8	(—) 0.96	I
39.83	20	?	.83	5	60 II	746.6	(w) 1.40 A^1	I
47.64	40	3	.65	6	80 II	698.8	(0.00) 0.84	I
64.99	2	2	5.02	1		593.3		II
77.36	100	5	.39	8	300 II	518.7	(0.00) 1.00	I
80.93	2	—	.93	1	5 III	497.3		I
81.20	3	—	.23	1	9 II	495.6		I
83.71	50	3	.71	5	100 II	480.6	(0.76) 0.53, **1.01,** 1.51	I
4095.39	1	—	—	—		410.8		
4102.38	150	10	.38	10	350 II	369.2	(0.00) 1.08	I
4103.79	2	—	—	—		24360.9		

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
4106.39	3	—	.39	1		24345.4		
10.80	3	—	.81	2	5 III	319.3		I
24.91	20	15	.93	5	20 III E	236.1	(0.00) 1.14	II
27.57	2	2	—	—		220.5		II
28.30	150	10	.32	10	300 II	216.2	(0.22) 1.15	I
42.84	100	6	.87	10	200 II	131.2	(0.00) 0.82	I
44.56	1	—	—	—		121.2		
57.63	5	—	.63	2	6 III	045.4		I
64.98	2	1	—	—		24003.0		
67.51	50	3	.52	8	100 II	23988.4	(—, 0.51, 0.85) 0.40, 0.74, 1.08, 1.40, 1.73	I
73.34	2	?	—	—		954.9		II
74.14	30	2	.14	4	100 II	950.3	(0.00) 1.10 A^1	I
77.54	100	125	.51	5	125 III E	930.9	(0.44) 0.90 B	II
90.06	3	—	—	—		859.3		
91.28	1	—	—	—		852.4		I
4199.27	10	5	.28	3	8 V E	806.9	(0.00) 0.63, 1.51, 2.41	II
4204.69	20	10	.70	4	20 V E	776.3	(0.00) 1.48	II
13.			.01	½	6 IV	729.3		I
13.53	2	—	.54	½	5 III ?	726.4		I
17.80	5	—	.79	2	10 III	702.4	(—) 1.06	I
20.62	10	1	.62	3	30 III	686.6	(w?) 1.03 B	I
24.25	3	—	.23	2	4 III	666.3	(—) 1.21	I
29.22	2	—	.18	2	2 III	638.5		I
32.53	2	—	.54	2	1 III	619.9		I
35.73	40	20	.71	8	40 IV E	602.1		II
35.94	20	—	—	—	100 II	600.9		I
37.12	(1)	E+H	—	—		594.3		I
49.87	1	—	—	—	1 III	523.5		I
50.37	1	—	.35	1	1 III	520.8		I
51.20	10	1	.18	5	40 III	516.2	(0.00) 0.96	I
54.35	1	—	—	—		498.8		I
56.43	1	—	—	—		487.3		
64.88	—	1h	—	—		440.7		II
66.89	1	—	.90	2	1 III	429.7		I
72.12	2	—	.13	2	2 III	401.0	(0.00) 1.14	I
74.16	1	—	.20	1	1 III A	389.7		I
79.3	2	5hl	—	—		361.8		II
91.03	2	—	.05	3	2 III	297.8	(—) 0.85	I
4296.66	2Ce?	—	—	—		267.4		
4300.34	2	—	.37	1	2 III	247.4	(—) 0.82	I
02.30	20	2	.30	6	50 III	236.9	(0.52 w) 1.04 A^2	I
07.70	1	—	—	—		207.7		
09.62	70	50	.61	8	125 III E	197.4	(w) 1.15 A^1	II
15.47	2	—	.49	3	2 III	165.9	(—) 1.43?	I
16.30	2	—	.30	3	3 III	161.5		I
17.87	1	—	—	—		153.1		I
18.21	1	—	—	—		151.2		I
22.29	1	—	.31	1	2 III	129.3	(0.00) 1.22	I
22.54	2	—	—	—		128.1		
24.57	1	—	.61	1		117.1		
29.89	1	—	—	—	1 III	088.8		I
30.78	5	—	.78	3	10 III	084.1	(0.00) 0.73	I
37.28	2	—	.32	1	2 III	049.4	(—) 1.13 B	I
44.64	3	—	.65	1	4 III	010.4	(—) 0.83 A^2	I
45.60	1	—	—	—		005.3		
46.13	1	—	.16	1	1 III	23002.5		I
48.78	25	?	.79	6	60 III	22988.5	(0.00) 1.10	I
52.33	2	—	.34	2	5 III	969.8	(—) 1.27	I
52.71	4	—	.65	2	4 III	967.9	(—) 0.96	I
4353.63	2	—	—	—		22962.9		I

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
4354.35	2	—	—	—		22959.1		
54.98	1	—	—	—		955.8		
57.73	6	—	.73	3	10 III	941.3	(—) 0.98	I
58.73	40	30	.72	5	50 III E	936.1	(0.94) 0.48, **1.46**	II
61.18	1	—	—	—		923.2		I
64.01	2	?	—	—		908.3		II
64.17	1	?	—	—		907.4		II
64.41	—	2h	—	—		906.2		
66.03	5	—	.04	3	10 III	897.7	(0.00) 1.18	I
70.96	1	—	.98	1	1 III	871.8	(—) 0.98	I
71.44	1	—	—	—		869.4		
71.78	1	—	—	—		867.6		
74.94	200	300	.95	8	300 III E	851.0	(0.15) 0.95	II
75.61	3	—	.60	1	8 III	847.6		I
78.59	1	—	—	—		832.0		
79.33	3	—	.35	2	6 III	828.1	(—) 1.19 A^1	I
84.80	2	—	—	—		799.7		I
85.48	2	—	.48	1	3 III	796.1	(—) 1.36	I
87.74	4	—	.75	3	8 III	784.4	(—) 0.63 A^1	I
94.01	2	—	.02	2	3 III	751.9	(—) 1.17 B	I
94.66	2	—	.68	2	2 III	748.5		I
97.79	(2)	E+H	—	—	3 III	732.3		I
4398.02	75	50	.03	6	100 III E	731.1	(**0.00**, 0.32) **0.79**, 1.10, 1.41	II
4401.13	2	—	—	—		715.1		I
04.85	3	—	—	—		695.9		I
09.3	1	—	—	—		673.0		
09.7	1	—	—	—		670.9		I
11.20	1	—	—	—		663.2		
15.37	2	—	.38	2	3 III	641.8	(w) 1.23 Bw	I
17.44	2	—	.45	1	2 III	631.2	(—) 1.22	I
21.74	1	—	—	—	1 IV	609.2		I
22.59	50	30	.60	8	80 III E	604.8	(0.00) 0.51	II
32.92	1	—	—	—	1 IV	552.2		I
36.13	1	—	.14	1		535.8		I
37.35	5	—	.33	3	10 III	529.7	(—) 0.90	I
43.66	7	—	.65	3	15 III	497.7	(—) 1.70 A^2	I
45.31	3	—	.30	1	3 IV	489.3	(—) 0.62	I
46.63	8	—	.64	3	15 III	482.6	(0.00) 1.32	I
59.01	1	—	—	—		420.2		I
65.27	3	—	.26	3	1 IV	388.8	(0.00) 0.92	I
65.4	—	10h	—	—		388.1		II
72.79	2	—	.77	1	2n IV	351.2		I
73.88	5	—	.89	3	7 III	345.7	(—) 1.05	I
75.73	10	1	.71	3	20 III	336.5	(0.50) 1.29 Bw	I
76.37	2	—	—	—		333.3		
76.95	10	1	.97	4	25 III	330.3	(—) 1.66 A^2	I
77.45	10	1	.46	3	25 III	327.9	(—) 0.00, 0.78, 1.54	I
78.99	1	—	—	—	2 IV	320.2		I
84.44	2	—	—	—	5 II ?	293.1	(—) 1.06	I
87.28	8	—	—	—	20 III	279.0	(w) 1.20 Bw	I
87.47	15	1	—	—	40 III	278.0	(—) 0.60	I
88.90	1	—	—	—		270.9		
91.75	3	—	—	—	5 III	256.8		I
4492.42	3	—	—	—	4 III	253.5	(—) 1.48	I
4505.95	25	2	.96	8	50 II	186.7	(0.00) 0.90	I
13.58	3	—	.58	2	4 III	149.3	(—) 1.24	I
14.02	5	—	.01	2	8 III	147.0	(—) 1.31	I
22.05	2	—	.05	½	3 III	107.7	(—) 1.79	I
27.24	40	2	.26	5	80 II	082.3	(0.00) 1.15 A^1	I
27.79	30	2	.81	3	50 II	079.6	(0.00) 1.05 A^1	I
28.10	2	—	—	—	3 III	078.1		I
33.50	1	—	—	—	1 IV	051.8		I
34.09	3	—	—	—	3 III	049.0		I
4536.31	1	—	—	—		22038.2		I

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
4537.16	2	—	—	—	2 III	22034.0		I
39.60	1	—	—	—	1 III	022.2		I
42.03	3	—	.04	1		22010.4	(—) 0.93	I
44.31	6	—	.32	3	10 III	21999.4		I
54.46	2	—	—	—	2 III	950.4		I
55.29	2	—	.32	2		946.3		I
59.37	7	—	.37	4	10 III	926.7	(—) **1.03**, 2.04	I
61.8	2n?	—	—	—		915.0		I
64.37	2	—	.41	1	4 III	902.6	(—) 0.91	I
64.93	2	—	.95	2	2 IV ?	900.0		I
70.65	2	—	.69	2	3n III	872.5		I
73.55	6	—	.58	4	10 III	858.7	(—) 1.20	I
78.87	1	—	—	—	1 III	833.3		I
81.32	2	—	.32	1	6 II A	821.7		I
81.77	1	—	.77	1	3 III A	819.5		I
82.15	1	—	.18	1	2 III	817.7		I
85.			.33	1		802.6		
90.			.80	3	1 V	776.6		I
94.			.00	1	15 II ?	761.4		I
4596.54	10	—	.56	5	12 III	749.4	(—) 1.34	I
4601.27	1	—	.30	1	2 III	727.0		I
03.7	2nl?	—	—	—		715.6		
04.79	6	—	.81	3	10 III	710.4	(—) 1.26	I
07.94	3	—				695.6		
13.00	2	—	.00	1	4 III	671.8		I
36.50	(1)	E+H	—	—	1 IV	562.0		I
43.69	50	5	.70	5	150 I	528.6	(0.00) 0.89	I
49.5	3nl	—	—	—		501.7		Mol
50.1	5nl	—	—	—		498.9		Mol
52.13	1	—	—	—		489.5		I
53.78	2	—	—	—	4 III	481.9		I
58.32	10	—	.32	3	12 III	461.0	(0.00) 0.94	I
58.88	3	—	—	—	6 III	458.4	(—) 1.00	I
66.38	1	—	—	—	3 III	423.9		I
66.84	2	—	—	—	3 III	421.8		I
67.47	4	—	—	—	8 III	418.9	(—) 0.84	I
70.82	2	—	—	—	4 III A	403.5	(—) 0.87	I
74.84	45	5	.84	5	125 I	381.5	(0.00) 1.06	I
78.36	2	—	.34	1	3 III	369.1		I
82.32	30	20	.31	5	20 V E	351.0	(0.89) 0.52, **0.95, 1.49**, 1.84	II
89.77	3	—	.76	1	4 III	317.1		I
91.			.97	1	1 III A	307.1		I
96.81	6	—	.80	2	8 III	285.1	(0.00) 0.97	I
4699.			.24	2	1 III ?	274.1		I
4701.00	1	—	.98	2	2 III	266.2		I
04.			.64	1		249.7		I
08.84	1+p	—	—	—	3 III ?	230.7		I
25.84	2	—	—	—	3 III	154.4		I
28.52	10	1	.53	4	20 III ?	142.4	(0.67) 1.29	I
32.35	3	—	.39	2	3 III ?	125.2		I
34.52	—	5h	—	—		115.6		II
41.40	5	—	.41	3	8 III	084.9	(0.00) 1.04	I
44.6	2nl	—	—	—		070.7		Mol
52.79	10	1	.78	3	12 III	21034.4	(0.00) 1.90 A^2w	I
60.98	30	3	.99	5	40 I	20998.2	(0.50, **0.80**) 0.40, 0.72, **1.04**, 1.36, 1.69	I
80.			.18	2	1 III	913.9		I
81.04	10	1	.04	3	10 III	910.1	(0.00) 1.45 A^2	I
86.58	30	20	.57	4	10 IV E	885.9	(0.65w) 1.20 Bw	II
86.88	10	—	.90	2	10 III	884.6		I
4799.30	15	2	.31	4	15 III	20830.5	(w) 1.52 A^2	I

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
4804.31	3	—	.31	3	4 III	20808.8		I
04.81	5	—	.81	3	6 III	806.7		I
17.38	5nl	—	.41	1		752.3		Mol
18.18	10nl	—	.20	1 BH		748.9		Mol
19.64	7	1	.64	3	10 III	742.7	(**0.31,** 0.95) 0.64, 1.28, **1.93**	I
21.62	1	—	.62	1	2 IV	734.1		I
22.12	6	1	.12	2	8 III	732.0	(w) 0.90	I
23.31	40	30	.32	4	12 V E	726.9	(0.46, **0.99**) 0.17, **0.64, 1.13,** 1.62	II
26.			.25	½		714.3		
29.36	—	10h	—	—		700.9		II
39.13	2	—	.16	1	3 III	659.3	(**0.34,** 0.96)—	I
39.87	60	6	.86	6	60 II	656.0	(0.00) 1.30	I
42.03	2nl ?	—	—	—		646.7		Mol
42.84	5nc ?	—	—	—		643.3		Mol
45.67	50	5	.68	4	50 II	631.2	(0.00) 1.20	I
52.69	40	4	.69	4	50 II	601.4	(0.00) 1.01	I
54.26	1	—	—	—	4 III	594.6		I
54.87	100	150	.88	10	70 V E	592.1	(0.00) 0.72	II
56.70	2	—	.70	2	3 III	584.4	(w?) 1.28	I
59.85	25	2	.83	4	40 II	571.1	(0.00) 0.40	I
63.			.11	2	1 III	557.2	(—) 0.80	I
64.			.71	1		550.5		
79.			.64	2	4 III	487.6	(0.70) 2.28	I
81.44	2	2	.44	2		480.0		II
83.69	150	200	.69	8	80 V E	470.6	(0.00) 1.11 A^1	II
86.29	3	—	.26	2	4 III	459.8	(0.00) 1.24	I
86.65	3	—	.64	2	3 III	458.2	(0.00) 0.84	I
4893.44	8	2	.44	3	6 III A	429.8	(**0.32,** 0.95) 0.70, 1.30, **1.90**	I
4900.13	125	150	.11	8	80 III E	402.0	(0.00) 0.97 A^1	II
06.11	8	1	.10	3	6 III A	377.1	(0.00) 1.50 A^3	I
09.00	2	—	.00	2	2 III	365.1	(0.00) 1.18	I
12.03	1	—	.07	2	tr III A ?	352.4		I
14.81	1	—	—	—		341.0		I
21.88	10	2	.85	3	10 III A	311.8	(0.35) **0.95,** 1.66	I
26.32	1	—	.30	1	tr III A	293.5		I
28.21	2	—	.24	2	1 III	285.6		I
30.93	3	—	.95	2	2 III	274.5	(0.00) 0.83	I
36.			.70	1		250.8		
48.54	1	—	—	—	tr IV	202.4		I
50.66	3	—	.63	2	2 III	193.8	(0.00) 0.96	I
70.			.10	1		114.7		
74.30	5	1	.31	3	10 III ?	097.7	(0.00) 1.08	I
79.			.24	1		077.8		
4982.13	15	15l	.12	5	8 V E	20066.2	(w) 1.56 A^3	II
5006.96	6	—	.97	3	10 IV	19966.6	(0.00) 0.87	I
24.3	2nl	—	—	—		897.7		Mol
25.2	3nl	—	—	—		894.2		Mol
49.7	2nl	—	—	—		797.7		Mol
50.6	2nl	—	.65	1		794.1		Mol
68.			.82	2		723.0		
70.21	2	—	.18	3	5 V	717.6		I
72.19	3	1	.19	2	5 IV	709.9		I
87.42	50	100l	.42	10	150 V E	650.9	(0.00) **1.22**	II
5088.18	1	—	—	—	3 V	647.9		I
5103.70	1	—	—	—		588.2		I
19.12	10	20l	.10	5	15 V E	529.2	(**0.00,** 0.40) 1.39, **1.92**	II
23.21	30	50l	.21	6	40 V E	513.6	(0.00) 0.52 A^1	II
28.42	1	—	—	—		493.8		I
35.20	4	2	.20	3	25 III	468.0	(0.00) 1.02	I
5196.43	5	10l	.43	3	2 V E	19238.6		II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
5200.42	40	60	.41	10	60 V E	19223.9	(0.00) 0.72	II
05.73	50	80	.71	10	80 V E	204.3	(0.00) 1.07	II
28.56	2	—	.50	1		120.5		
40.80	4	2	.79	3	15 IV	19075.8	(0.00) 0.94	I
83.69	2	—	.66	1		18921.0		I
5289.82	4	5l	.81	4	4 V E	899.0	(w) 1.73 A[2]	II
5320.78	2	4	.77	2		789.0		II
25.84	1	—	—	—		771.2		I
27.12	1	—	—	—		766.7		
75.87	1	—	—	—		596.5		I
80.63	5	1	.61	3	15 III	580.1		I
88.40	1	—	.42	1		553.2		I
5390.82	2	—	.79	1		545.0		I
5402.78	20	50l	.78	6	20 V E	503.9	(0.00) 0.89	II
17.02	2	—	.03	2	3 IV ?	455.2		I
24.36	5	1	.37	3	20 III	430.2	(0, 22, 0.66) 1.78 A[2]	I
38.24	20	2	.24	5	50 II	383.2	(0.00) 1.09 B?	I
66.46	50	10	.45	10	300 II	288.3	(0.00) 1.10	I
68.47	10	1	.46	4	40 III	281.6	(0.39) 1.04	I
69.19	1	—	—	—		279.2		I
73.40	10	20l	.38	5	10 V E	265.2	(0.00) 1.48	II
80.75	10	15l	.72	5	8 V E	240.7	(0.00) 1.48	II
91.46	—	2h	.44	3	1 V	205.1		I
93.18	5	1	.15	3	15 III	199.4	(0.22) 0.62	I
95.61	2	—	.57	3	8 III	191.4		I
5497.42	20	50	.41	5	25 V E	185.3	(0.00) 1.43	II
5501.			.52	1		171.8		
03.34	2	—	—	—	100{5 III / 25 II}	165.8		
03.47	10	2	.43	5		165.4	(0.00) 1.15	I
09.91	30	30l	.90	6	40 V E	144.1	(0.45) 0.76 Bw	II
13.66	5	1	.65	2	20 II	131.8	(0.65)—	I
21.59	10	—	.60	5	} 30 II+E{	105.7		I
21.70	2	20	—	—		105.3	(0.00) 1.46	II
26.75	2	—	.72	2	2 IV	088.8		I
27.56	40	4	.53	8	250 II	086.2	(0.00) 1.03 A[1]	I
27.75	10	—	—	—		085.5		I
40.			.61	1		043.6		I
41.64	2	—	.63	2	3 III	040.2		I
44.61	15	10l	.60	5	25 III+E	030.5	(0.00) 1.50	II
46.02	8	10l	.02	4	3 V E	18026.0	(0.00) 1.46	II
56.44	5	1	.42	5	20 I A	17992.2	(0.00) 0.90	I
67.76	5	1	.74	2	15 II	955.6		I
77.43	10	2	.42	3	30 II	924.5	(0.00) 1.06	I
81.08	2	—	.07	1	3 IV	912.7		I
81.88	30	4	.86	5	150 II	910.2	(0.00) 0.92	I
90.96	2	—	.95	2	2 IV	881.1		I
5594.12	2	—	—	—	2 IV	871.0		I
5603.			.26	1		841.8		I
06.34	10	1	.32	3	20 II	832.0	(0.62) 1.20 Bw	I
10.36	1	2	.34	1		819.3		II
19.98	1	—	—	—		788.7		I
23.89	2	—	.90	1		776.3		I
30.15	20	2	.12	6	100 II	756.6	(0.00) 0.71 A[2]	I
32.25	2	—	.23	1	3 I A	750.0		I
32.92	1	—	.86	1		748.0		I
44.70	10	1	.68	4	20 II	710.9	(0.85) 1.10 Bw	I
46.70	1	—	.66	1		704.6		I
48.48	10	1	.45	4	20 II	699.0	(0.00) 1.06	I
60.			.90	1		660.1		I
5662.95	50	200	.95	8	50 V E	17653.8	(0.00) 0.98	II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
5669.24	1	—	.19	1	1 IV	17634.2		I
75.29	5	1	.26	3	15 II	615.4	(1.14) 0.82 Bw	I
75.63	1	—	—	—	1 V	614.3		I
86.			.62	1	1 IV ?	580.3		I
5693.63	2	—	.62	1	2 II A	558.6		I
5706.75	15	1	.72	4	30 II	518.3	(0.00) 1.12	I
13.81	1h	—	—	—	3n IV	496.6		I
14.91	1h	—	—	—	3n IV	493.2		I
20.62	2	—	.60	2	3 IV	475.8	(1.23) 1.32	I
23.46	1	—	.43	2	1 IV	467.2		I
28.91	10	10l	.89	4	3 V E	450.5	(0.80w) 1.27 Bw	II
32.10	2	—	.08	1	3 II	440.8	(—) 1.35	I
40.22	1	—	—	—	1 III A	416.1		I
43.87	6	1	.85	3	15 III	405.1	(0.00) 1.28	I
63.58	1	—	—	—		345.5		I
64.22	1	—	—	—		343.6		
65.67	5	1	.63	3	12 III	339.3	(0.00) 1.18	I
73.95	2	—	.93	1	4 IV	314.4	(0.62) 0.48	I
81.69	5	5l	.68	4	2 V E	291.2	(0, 66, 0.41) 0.73 A[1]	II
82.68	1	—				288.2		
87.70	1	—	.72	1	1 IV	273.2		I
5797.15	1	—	.16	1	1 IV A	245.2		I
5812.64	2	—	.692	1		199.1		I
18.58	2	—	—	—		181.6		
21.87	3	—	.891	2	4 III	171.8		I
32.25	2	—	.275	2	3 IV	141.3		I
44.			.60	1		105.1		
58.82	1n?l	—	—	—		063.6		
71.80	2	—	.851	1	2 IV	025.8		I
76.13	1	—	—	—		013.3		
79.94	2	—	.971	2	3 IV	17002.2		I
93.94	1h	—	4.043	1u		16961.8		
5895.			.89	1		956.3		
5902.93	4	—	.979	3	7 IV	936.0	(—)1.50	I
12.18	1hn?	—	.20	1BH		909.5		Mol
31.09	1nl	—	.12	1BH		855.6		Mol
39.06	5nl	—	.035	2BH		833.1		Mol
44.			.871	2		816.6		I
45.71	4	—	.729	4	6 IV	814.2	(0.00) 1.52 B?	I
50.00	3	—	.036	3	5 IV	802.0	(0.64)—	I
56.40	4nl	—	.383	1BH		784.1		
66.			.642	2		755.2		
72.1	100nl	5nl	.05	6BH		740.0		Mol
81.86	3	—	.920	3BH		712.5		I
87.6	80nl	4nl	.640	5BH		696.5		Mol
5992.12	2	—	—	—		684.0		
6003.6	60nl	3nl	.583	4BH		652.1		Mol
04.65	5	—	—	—		649.2		
07.64	2	—	.742	3		640.7	(0.00) 1.24	I
09.16	8	1	.227	5	10? IV	636.6	(0.00) 1.40	I
19.9	40nl	2nl	.87	4BH		607.0		Mol
23.40	4	—	.426	3	20? I A	597.3	(--) 0.60, 0.98, 1.36	I
24.26	1p?	—	.330	2u	? III	594.9		I
36.6	30nl	2nl	.600	3BH		561.0		Mol
40.			.283	1		550.9		I
53.8	20nl	1nl	.785	3BH		514.0		Mol
60.			.34	1		496.2		
72.8	5nl	—	.84	2BH		462.3		Mol
81.			.221	1		439.5		
87.99	2	—	8.010	3		412.2	(0.00) 1.52	I
89.4	10nl	—	.37	2BH		417.5		Mol
6096.8	8nl	—	.77	2BH		397.5		Mol
6107.8	5nl	—	.85	2BH		367.9		Mol
14.7	6nl	—	.760	2BH		349.4		Mol
6122.			.192	2		16329.5		

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter		Zeeman effect	Spectrum
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν		
	Arc	Spark						
6127.39	2p	—	.43	1BH		16315.6		Mol
32.1	80nl	6nl	.08	5BH		303.2		Mol
35.04	2	—	.056	2	? III	295.3	(0.00) 1.50	I
38.42	5	—	.456	3	15? I A	286.3	(0.39) 1.28B	I
48.4	70nl	5nl	.38	4BH		259.9		Mol
51.72	2	—	—	—		251.1		
62.			.203	1		223.5		
65.1	60nl	4nl	.059	4BH		216.0		Mol
82.2	50nl	3nl	.233	4BH		171.2		Mol
91.73	50	5	.726	6	100? I A	146.1	(0.00) 0.78	I
6199.8	40nl	2nl	.818	3BH		125.1		Mol
6217.9	20nl	1nl	.929	3BH		078.1		Mol
22.59	10	2	.585	5	50? I A	066.0	**(0.12,** 0.58) 1.75A²	I
36.7	10nl	—	.690	2BH		16029.7		Mol
51.06	3	—	.045	2		15992.9		I
65.			.08	1		957.1		
75.01	5nl	—	.060	2BH		931.8		Mol
6295.46	3nl	—	.447	1BH		880.1		Mol
6316.20	2nl	—	.367	1BH		827.7		Mol
32.			.23	1		787.9		
38.12	3	—	.150	1	5 I	773.1		I
45.			.98	1		753.7		
57.38	1	—	—	—		725.4		
69.87	1	—	—	—		694.6		
87.08	2	—	—	—		652.3		
6396.			.36	1		629.6		
6402.01	10	2	.025	4	50 I A	615.8	**(0.21,** 0.60) 1.76A²	I
05.59	1	—	—	—		607.1		
35.02	100	20	.030	7	500 I A	535.7	(0.00) 1.16	I
37.17	3	—	.200	2	5 IV ?	530.5		I
62.59	2	—	.58	1	8 III	469.4		I
82.6	1	—	—	—		421.7		
6493.8	1	—	.77	1	4 V	395.1		I
6501.3	2	—				377.3		I
05.			.44	1		367.5		
18.35	2	—	—	—		337.1		
35.88	2	—	—	—		295.9		
38.59	15	4	.599	4	35 III	289.6	(0.00) 1.12	I
53.88	2	—	—	—		253.9		
57.38	10	2	.435	3	30 I A	245.7	(0.00) 1.37A².	I
72.6	2	—	—	—		210.5		
76.86	5	1	.889	2	6 III	200.6	(—) 1.40	I
84.88	2	—	.898	1	5 II A	182.1		I
6595.04	1	—	—	—		158.7		
6602.			.40	1		141.8		
03.			.35	1		139.7		
13.74	25	20	.76	6	15 V E	115.8	(0.00) 1.24W	II
16.			.59	1		109.4		I
22.5	1	—	.50	1	2 IV ?	095.9		I
36.48	2	—	.49	3	2 IV ?	064.1		I
50.60	2	—	.60	2	8 IV	032.1	(0.20) 1.04	I
64.40	2	1	.37	4	3 IV ?	15001.0		I
83.			.26	1		14958.6		
87.57	25	5	.57	6	80 I	949.0	(0.52) 0.63B	I
91.84	2	1	.81	1	2 V	939.5		I
94.75	1	—	—	—	1 V	933.0		I
6699.26	2	1	.32	2	3 IV	922.8		I
6700.71	15	4	.71	5	20 III	919.7	(0.00) 0.89	I
13.19	4	2	.21	3	6 III	891.9		I
35.98	5	2	.99	4	7 III	841.6		I
50.			.26	1		810.2		
54.			.96	2		799.9		
61.			.50	1		785.5		
62.			.16	3		784.1		
6777.			.19	1		14751.3		

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
6785.			.19	2		14733.9		
93.71	25	6	.72	4	80 I	715.4	(—) 1.10B	I
6795.41	20	30	.41	5	15 V E	711.8	(—) 0.83	II
6803.16	1	—	.12	1	1 V	695.0		I
15.16	2	2	.18	2	5 IV	669.1		II
32.49	3	4	.51	3	1 V E	631.9		II
45.24	5	2	.24	4	10 III	604.7	(—) 1.18	I
58.25	3	5	.22	1	1 V E	577.0		II
87.22	2	4	.24	4	8 IV	515.6	(—) 0.99	I
6896.00	5	10	.00	2	3 V E	497.2	(—) 2.21	II
6906.			.34	2		475.5		
08.26	2	—	.36	2	2 III	471.3		I
29.			.17	2		427.8		
33.55	2	2	.56	3	6 II A	418.7		I
50.30	4	4	.34	4	8 IV	383.9		I
51.68	1	3	.69	4	1 V E	381.1		II
53.			.37	3		377.6		
58.04	2	1	.11	2	4 I A	367.8		I
6979.88	4	2	.89	4	8 V ?	323.0		I
7008.95	3	2	.99	3	5 V	263.5		I
09.93	2	2	.93	2	4 V	261.6		I
35.18	3	2	.18	3	5 IV	210.3		I
52.95	4	3	.93	1	10 III	174.6		I
54.28	3	1	—	—	4 V	171.9		I
7075.17	2	1	—	—		130.0		I
7127.92	3h?	1	8.10	2	2 III ?	025.5		I
39.			.74	2		14002.4		
55.			.40	1		13971.6		
91.65	5	2	.68	4	10 III	901.2		I
93.74	1	1	—	—		897.2		II
7195.95	2	1	.94	2		892.9		I
7264.19	7	10	.14	5	8 IV E	762.4		II
7293.10	2	—	—	—		707.8		I
7303.2	1	—	—	—		688.9		I
30.62	2	—	—	—		637.3		
32.97	1	2	—	—		633.3		II
46.47	10	2	.28	4	10 III	608.2		I
88.46	1	1	—	—		530.9		II
7398.80	4	1	—	—		512.0		I
7406.23	1	2	—	—		498.4		II
50.32	5	5	.21	5	5 III ?E	418.5		II
55.20	2	—	—	—		409.8		I
72.2	2	—	—	—		379.3		I
78.8	2	—	—	—		367.5		I
86.4	1	—	—	—		353.9		I
7494.90	5	1	—	—		338.7		I
7526.0	1d	—	—	—		283.6		
36.73	3	1	—	—		264.2		I
53.28	1	—	—	—		235.6		
7563.13	10	4	—	—	5 V	218.4		I
7617.72	4	—	—	—		123.7		I
22.94	5	1	—	—		114.7		I
52.89	3	—	—	—		063.4		I
89.49	2	—	—	—		13001.2		I
7698.00	4	—	—	—		12986.8		I
7719.89	6	1	—	—	2 V	950.0		I
24.08	5	1	—	—	2 IV	943.0		I
88.42	3	—	—	—		836.1		I
7796.32	4	—	—	—		823.1		I
7802.52	2	—	—	—		812.9		I
12.16	5	1	—	—		797.0		I
23.94	1	—	—	—		777.8		I
55.52	7	1	—	—		726.4		I
70.04	2	—	—	—		702.9		I
7881.90	20	10	.69	2	10 V E	12683.8		II

TABLE 1.—*Wave lengths and Zeeman effects in yttrium spectra*—Continued

Meggers			Eder		King and Carter			
λ	Intensity		λ	Intensity arc	Arc intensity and temperature class	ν	Zeeman effect	Spectrum
	Arc	Spark						
7887.51	2		—	—		12675.0		I
7984.8	1		—	—		520.4		
7999.33	3		—	—		497.6		I
8025.60	3		—	—		456.7		I
8066.20	3		—	—		394.0		II
8134.9	3		—	—		289.3		I
65.5	2		—	—		243.3		I
8194.8	3Na?		—	—		199.5		
8211.71	4		—	—		174.4		I
31.23	2		—	—		145.5		I
47.4	1		—	—		121.7		
58.5	2		—	—		105.4		I
8297.07	2		—	—		049.2		I
8326.40	1		—	—		006.7		I
29.61	5		—	—		002.1		I
30.92	1		—	—		12000.2		I
44.43	10		—	—		11980.8		I
8365.64	4		—	—		950.4		I
8443.28	1		—	—		840.5		I
50.36	8		—	—		830.6		I
8475.64	3		—	—		795.3		I
8528.94	4		—	—		721.6		I
52.42	1		—	—		689.4		I
56.04	2d?		—	—		679.0		
75.77	2		—	—		657.6		I
8595.8	1		—	—		630.4		
8627.9	1		—	—		587.2		I
8658.4	1		—	—		546.3		I
8702.1	1?		—	—		488.3		
8759.24	1		—	—		413.4		I
8800.62	10		—	—		359.7		I
31.2	1		—	—		320.4		I
35.85	2		—	—		314.4		II
8876.6	1?		—	—		11262.5		
9231.58	8		—	—		10829.4		I
9392.7	1		—	—		643.6		I
9494.81	2		—	—		10529.2		I

WASHINGTON, May 25, 1928.

WEAR AND MECHANICAL PROPERTIES OF RAILROAD BEARING BRONZES AT DIFFERENT TEMPERATURES

By H. J. French, S. J. Rosenberg, W. LeC. Harbaugh, and H. C. Cross

ABSTRACT

A study was made of the wearing and mechanical properties of two groups of copper-tin-lead alloys in both chill-cast and sand-cast conditions. The bronzes were tested under rolling and sliding friction without lubrication at atmospheric and elevated temperatures up to 350° F. (175° C.), under both sliding and rolling and sliding friction with lubrication at atmospheric temperatures, and under tension, impact, and pounding at temperatures from 70 to 600° F. (20 to 315° C.). A detatiled description is given of all apparatus used in the tests.

No one of the selected laboratory tests yielded information which by itself was adequate for general comparisons.

Variations in chemical composition of the bronzes produced major changes in properties, but within certain ranges variations in chemical composition were much less important than variations in the method of casting.

Generally, chill-cast bronzes were less resistant to wear and impact than sand-cast bronzes but were more resistant to pounding and tension.

Increase in lead, particularly in the range 0.25 to 10 per cent, improved wearing properties but lowered resistance to pounding, impact, and tension. Increase in tin improved all properties. Increase in temperature of test generally decreased resistance to tension, impact, pounding, and wear, the changes being most marked from 350 to 600° F. (175 to 315° C.).

CONTENTS

	Page
I. Introduction	344
II. The problem of testing bearing metals	348
III. Previous investigations	349
IV. Alloys tested and their preparation	353
V. Equipment and test methods employed	357
1. The Amsler wear testing machine	357
(*a*) Calculation of unit contact pressures	361
2. Sand abrasion tests	362
3. Single-blow impact tests	363
4. Tension tests	365
5. Repeated-blow pounding tests	365
VI. Experimental results	366
1. Wear under sliding and rolling friction in the presence of oil	366
2. Wear tests at ordinary temperatures in the absence of oil	371
(*a*) Film formation	372
(*b*) Bronzes varying in lead and in tin	376
(*c*) Data showing that the wear test results can be duplicated	382
3. Wear tests at elevated tempreatures in the absence of oil	384
(*a*) Bronzes varying in lead	384
(*b*) Bronzes varying in tin	387
(*c*) Frictional properties *v.* wear	389
(*d*) Wear of the steels	392

109673°—28——1

VI. Experimental results—Continued.
3. Wear tests at elevated temperatures in the absence of oil—Continued. Page
4. Sand abrasion tests 393
5. Single-blow impact tests 394
6. Tensile tests 398
7. Repeated-blow pounding tests 406
VII. General discussion 410
1. Questions relating to methods of test 410
2. Chill casting *v.* sand casting 416
3. Effects of lead and tin in bearing bronzes 417
4. Specifications 418
VIII. Summary and conclusions 419

I. INTRODUCTION

The need for a study of bearing bronzes was recognized about 1920 by certain of the executives of the Bostwick-Lyon Bronze Co., Waynesboro, Pa., who felt that such wide variations encountered in the specifications of different carriers for bearing metals were against the interests of the carriers as well as those of the metal manufacturers.

During the latter part of 1925, subsequent to the merger of the interests of the Bostwick-Lyon Bronze Co. with those of the Chicago Bearing Metal Co., Chicago, Ill., work was begun under the research associate plan [1] at the Bureau of Standards. During the summer of 1927, while the tests were still in progress, the interests of the Chicago Bearing Metal Co., were merged with those of the Magnus Co., Chicago, Ill., which continued to support the work until September, 1927. The final phases of the program, including most of the mechanical tests, were completed subsequent to this time by the staff of the Bureau of Standards.

The tests were started largely through the active efforts of E. R. Darby, formerly chief metallurgist, and W. M. Corse, formerly consulting metallurgist, Chicago Bearing Metal Co., who acted in an advisory capacity throughout the investigation. Cooperation with the Magnus Co. was largely through R. J. Shoemaker, engineer of tests.

Between 1920 and 1925 the technical staff of the Chicago Bearing Metal Co. developed certain definite views concerning the manufacture and practical application of bearing bronzes. A search of the literature for support or disproval of these views revealed mostly theories and personal opinions, with few dependable facts. Records of service tests of bearings were conflicting. Likewise, specifications found to be satisfactory by one group seemed to fail in the hands of another.

[1] Described in B. S. Circular No. 296.

LOCOMOTIVE BEARINGS

Progress toward standardization of specifications for bearing metals was made by some of the technical societies, but due to lack of definite reproducible data, or for other reasons, many individuals in these organizations refused to adopt recommendations. While a natural tendency toward standardization could be detected during this period,

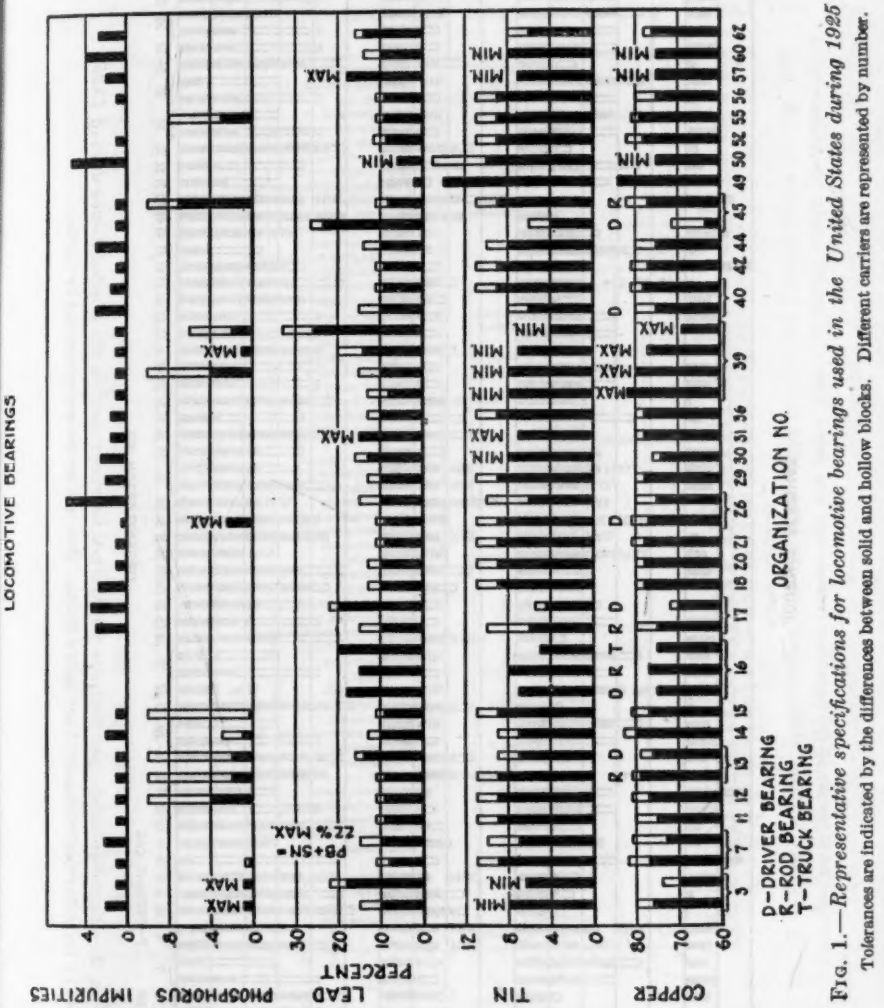

FIG. 1.—*Representative specifications for locomotive bearings used in the United States during 1925*

Tolerances are indicated by the differences between solid and hollow blocks. Different carriers are represented by number.

there was still a wide variation in specifications for similar types of bearings at the end of the year 1925, as will be evident from Figures 1, 2, and 3.

In January, 1925, an important trunk-line railroad experienced considerable trouble with "back-end main-rod brasses" which were breaking in service and causing costly delays along its lines. Consideration was given to replacement of sand-cast alloys by alloys cast

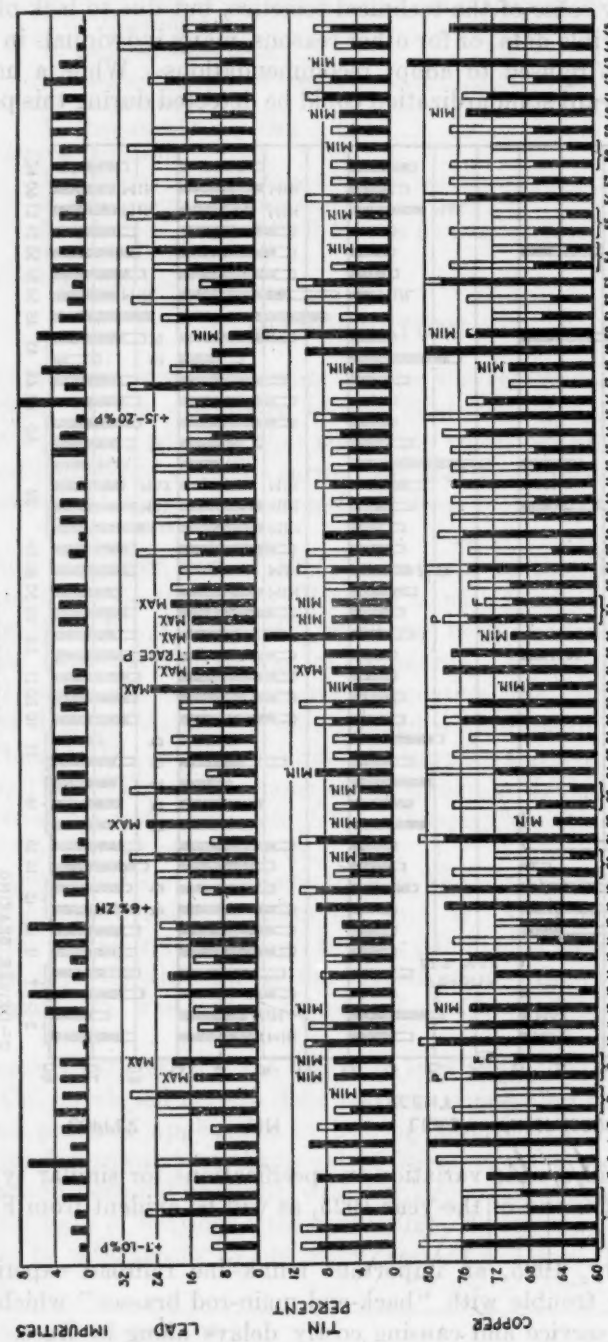

FIG. 2.—*Representative specifications for journal bearings used in the United States during 1925*

Tolerances are indicated by the difference between solid and hollow blocks. Different carriers are represented by number.

in metal molds in an attempt to overcome these difficulties, but an investigation developed conflicting views and little information on which to base predictions that this and other proposed changes would be beneficial.

The lack of such information was probably the most important single factor which influenced the Chicago Bearing Metal Co. to

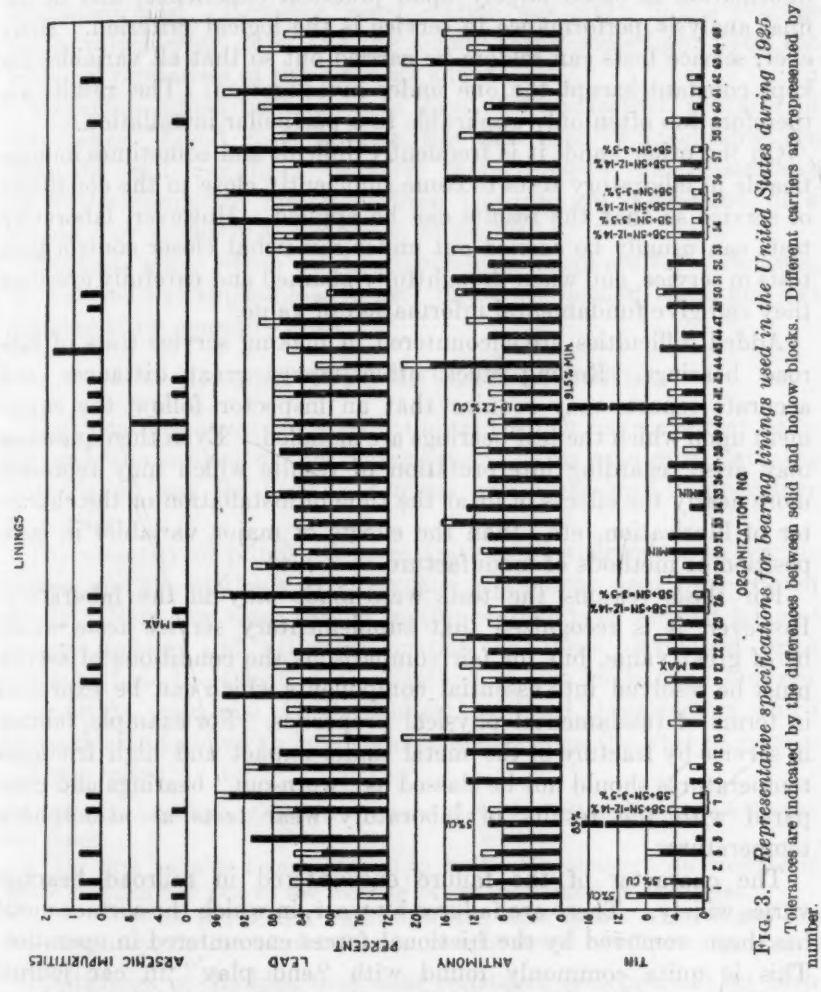

FIG. 3.—*Representative specifications for bearing linings used in the United States during 1925*
Tolerances are indicated by the differences between solid and hollow blocks. Different carriers are represented by number.

undertake the investigation here reported. The primary object of the tests was to determine important physical properties of a group of copper-tin-lead alloys within, and in some cases beyond, the limits of chemical composition and constitution ordinarily applied industrially in railroad bearings and to supply information that might ultimately permit a logical revision of specifications.

II. THE PROBLEM OF TESTING BEARING METALS

The first problem encountered in the study of bearing metals is the choice of test methods. At present there are no generally accepted methods of test nor any well-defined procedure which may be followed in attempting comparisons of different bearing metals. Reliable information is based largely upon practical experience, and in the final analysis performance in service is the logical criterion. However, service tests can seldom be carried out so that all variables are kept constant except the one under investigation. The results are therefore too often only applicable to a particular installation.

On the other hand, it is frequently difficult and sometimes impracticable in laboratory tests to come sufficiently close to the conditions of service so that the results can be applied. However, laboratory tests can usually be carried out under somewhat closer control than tests in service, and when thoughtfully planned and carefully executed they can give fundamental information of value.

Added difficulties are encountered in making service tests of railroad bearings. Rolling stock often moves great distances, and accurate reports may require that an inspector follow the equipment upon which the test bearings are installed. Even then questions may arise regarding interpretation of results which may represent more nearly the effects of fit at the time of installation or the character of lubrication, etc., than the effects of major variables in composition or methods of manufacture.

For these reasons the tests were made only in the laboratory. However, it is recognized that supplementary service tests would be of great value, but for fair comparisons the conditions of service must be resolved into essential components which can be expressed in terms of fundamental physical properties. For example, failures in service by fracture of the metal under impact and high frictional temperatures should not be classed as "worn-out" bearings and compared with the results of laboratory wear tests at atmospheric temperatures.

The character of the failure encountered in railroad bearings varies widely. There are failures by wear, in which the surface metal has been removed by the frictional forces encountered in operation. This is quite commonly found with "end play" in car journal bearings. Another type of failure is that due to deformation or "pounding out." A third type of failure commonly observed is fracture of the bearing.

Service failures of such types indicate that consideration should be given in laboratory tests of railroad bearing metals to (1) the resistance to wear and (2) resistance to both static and dynamic stresses. Since railroad bearings are frequently called upon to operate for

short periods at elevated temperatures and some types ordinarily "run hot" (at temperatures as high as 500° F. (260° C.) or more), the different properties should be considered at elevated temperatures as well as atmospheric.

There are other items which should be considered. One is starting torque, but it was not practicable to study this feature with the equipment used. However, consideration was given to the average torque in the wear tests, and the results throw some light upon the frictional properties of the different alloys.

Properties which contribute to ease of manufacture are also of importance. Good casting properties and ease of machining are desirable, and variations in such characteristics may materially affect the choice of metals. These features are considered in a general way in the final discussion of results.

The tests made which will be described in detail are (1) wear tests, (2) notched-bar single-blow impact tests, (3) tension tests, and (4) repeated-blow pounding tests.

III. PREVIOUS INVESTIGATIONS

Technical literature contains little data which have a direct bearing upon the subject of this investigation. This applies to both the methods of test for bearing bronzes and their high temperature properties.

The majority of published investigations of wear relate to ferrous metals, and the methods of study and equipment used have varied widely. Very few wear testing machines have been applied to nonferrous metals, and little information has been developed upon which to base the selection of a method of test for bearing bronzes.

About 35 years ago, C. B. Dudley [2] outlined three factors which he considered to have an important influence upon the wear of metals. These were:

1. That metal which will suffer the most distortion without rupture will wear best (least).

2. The first variable being obtained in satisfactory amount, an increase in tensile strength will add to the wearing power (resistance) of the metal.

3. * * * of two metals which have the same tensile strength and the same elongation, the one which is finer in granular structure will wear the slower.

This brings in the idea that wear resistance is dependent on both hardness and toughness, and this has been confirmed in more recent work.

Portevin and Nusbaumer,[3] using a Derihon "abrasion mill," tested a number of copper-tin alloys with and without additions of

[2] C. B. Dudley, Bearing metal alloys, J. Franklin Inst., **133**, pp. 81 and 161; 1892.

[3] Portevin, A., Nusbaumer, E., "Note on the wear of bronzes," Intl. Assn. of Test. Matls., 6th Congress, III; 1912

phosphorus and zinc. They reported that "when the bronzes are submitted to constant friction, account should be taken of the phenomena of cold work, which tends to diminish the wear until the limit of the depth to which the cold work has penetrated is reached." They also concluded that the introduction of phosphorus decreased the wear in high-tin bronzes and increased the wear in low-tin bronzes and that the wear (weight loss) of their copper-tin alloys was "proportional to the tin percentage or, more exactly, to the percentage of delta (constituent)."

Jannin[4] made wear tests under lubrication and arrived at the conclusion that the principal cause of wear of bearing metals was insufficient polish of the bearing and that the principal cause of wear of the shaft was lack of cleanliness of the lubricant. He also found that leaded bronzes were at first less resistant, and later more resistant, to wear than bronzes practically free from lead. His phosphorus bronzes showed a relatively high resistance to wear.

G. H. Clamer[5] reported some of the general conclusions drawn from tests made for a thesis by Messrs. Knapp and Miskella at Cornell University about the year 1903. Clamer's summary of this work follows:

1. Operating temperature of a bearing is of importance and should, therefore, be properly considered when selection of alloys for car journal bearings is to be made. A bearing which operates normally at a low temperature is less liable to reach the danger point resulting in so-called hot box when the ideal conditions of lubrication are disturbed.

2. Friction and wear are two factors which depend upon the lubrication of the bearings. * * * The property of a bearing to absorb oil is of considerable importance.

3. Wear on both journal and bearings is regulated to a large extent by the film of lubrication existing between the two rubbing surfaces.

4. * * * bearing tests which record the action of one metal as against another, as far as friction and wear are concerned, are more or less unsatisfactory, due to the element of lubrication, which of itself at the best must be more or less imperfect. If by some method the influence of this element could be removed, a better idea of the wearing qualities of the metals tested comparatively would be secured.

As a result of further study of copper-tin-lead alloys and also copper-tin-lead-zinc alloys, Clamer supported these conclusions and made a number of recommendations concerning car journal bearings. These were that "65 per cent of copper is the lowest percentage of copper which should be used * * *; there should be a proper balancing of the four metals used—that is, copper, tin, lead, and zinc"—but that it did not seem desirable in car journal bearings to have over approximately 5 per cent of tin.

[4] Jannin, M. L., "Nouvelle methode d'essai des metaux a l'usure, Machine d'essai a l'usure," Rev. de Met. Mem., **19**, p. 109; 1922.

[5] Clamer, G. H., "Effect of changes in composition of alloys used by American railroads for car journal bearings," Tr. Am. Inst. Metals, **9**, p. 241; 1915.

Clamer's wear tests showed that an increase in zinc tended to increase the rate of wear in bronzes and that it was desirable to increase lead and decrease the tin in the range of compositions studied. As stated by Clamer, "a better wearing (more resistant) metal results from an alloy of 70 per cent copper, 5 per cent tin, 20 per cent lead, and 5 per cent zinc than with an alloy of 80 per cent copper, 10 per cent tin, and 10 per cent lead."

Joachim and Case[6] studied the wear of a variety of commercial bearing materials, including copper-tin-lead alloys. The different materials were tested in the presence of gasoline, and it was found that when a machine-steel shaft was used a hard bearing material wore less below the seizing load than a soft bearing material. This order was reversed when the bearing material was operated in contact with a hardened tool-steel shaft. In general, the hard bearing materials produced more wear on the shaft than the soft bearing materials.

Kuhnel and coworkers[7] recently studied the copper corner of the copper-tin-zinc system. An alloy containing about 86.5 per cent copper, 8 per cent tin, 3.5 per cent zinc, 2 per cent lead, and small amounts of nickel and iron was less resistant to wear than a lower tin alloy, containing 83 per cent copper, 6 per cent tin, 9 per cent zinc, 2 per cent lead, and small amounts of nickel and iron.

The mechanical properties of copper-tin alloys at ordinary temperatures have received considerable attention. Increase in tin from 0 to about 19 per cent raises the tensile strength from around 25,000 lbs./in.^{2} to about 40,000 lbs/in.^{2} At the same time hardness increases,[8] and under equilibrium conditions the structure changes from (1) a single phase comprising the alpha solid solution of tin in copper to (2) a duplex or two-phase condition. In (1) all crystals have similar properties, but in (2) particles of the hard alpha-delta eutectoid are embedded in the relatively soft alpha matrix.

According to the results reported by Matsuda,[9] increase in tin raises the resistance of bronzes to repeated impact in the Stanton machine until appreciable proportions of the alpha-delta eutectoid appear in the structure, when the number of blows required for fracture of the test specimens decreases materially and continues to decrease with further increase in the tin.

According to Hoyt,[10] the copper-tin alloys containing a considerable amount of the eutectoid possess the desirable properties of

[6] Joachim, W. F., Case, H. W., "Tests of several bearing metals lubricated by gasoline," Technological Notes, No. 241, Natl. Adv. Com. for Aeronautics; 1926.

[7] Kuhnel, R., "On the constitution and properties of red brass," Zeit. f. Met., **18**, pp. 306-11; 1926; J. Inst. Met., No. 1, p. 465; 1927.

[8] Matsuda, T., Shiba, J., "On the hardness of brass, aluminum bronze," and bronze, Sci. Rpts. Tohoku Imp. Univ., **13**, No. 4, p. 413; 1925.

[9] Matsuda, T., "Repeated impact tests on brass, aluminum bronze, and bronze," Sci. Rpts. Tohoku Imp. Univ., **13**, No. 4, p. 419; 1925.

[10] Hoyt, S. L., Metallography—Metals and Common Alloys, **2**, McGraw-Hill Co., New York, p. 116 et seq., 1922.

mechanical strength and low coefficients of friction but lack sufficient plasticity and are too liable to overheat locally to make good bearing alloys. Additions of lead in proportions from 5 to 30 per cent improve plasticity but lower the strength.

Much less attention has been given to the properties of copper-tin alloys and the leaded bronzes at elevated temperatures. A summary prepared by Upthegrove and White in 1924 [11] will serve to illustrate the general effects of temperature on the mechanical properties of these alloys:

> The influence of temperature upon the properties of copper-tin bronzes, according to Bregowsky and Spring,[12] becomes most marked between 400° and 600° F. (205° and 315° C.), all of the properties except the elastic limit decreasing very rapidly within that range of temperature. At temperatures below 300° to 350° F. (150° to 175° C.) no appreciable difference in tensile strength is found for the bronzes carrying 12 per cent of tin. At all higher temperatures the 12 per cent bronze is superior in strength to the bronze of a lower tin content. The influence of the additional tin is also shown in the effect on elongation and reduction of area at temperatures above 750° F. (400° C.). Variations in properties at elevated temperatures seemingly are dependent upon the presence of the delta constituent, the final decrease in elongation and tensile strength coming at the temperatures corresponding to the absorption of this constituent.
>
> Low-tin bronze, according to Huntington,[13] shows a gradual decrease in strength with increasing temperature. Elongation and reduction of area decrease rapidly at 500° F. (260° C.), reaching a minimum at 700° F. (370° C.). The results obtained by Huntington are very similar to the results obtained in 1893 by Rudeloff [14,15] with copper carrying 1.86 per cent tin. This copper-tin alloy shows a tensile strength superior to that of copper at all temperatures up to 900° F. (480° C.).

Matsuda [16] studied dynamic hardness of bronzes at different temperatures. The dynamic hardness number represents the energy required to depress a unit volume of metal when a weight having a spherical penetrator similar to a Brinell ball is dropped on the specimen. Alpha bronzes showed a regular and gradual decrease in dynamic hardness with increase in temperature, while bronzes containing a considerable proportion of the eutectoid showed a sharp decrease within the transformation temperature range. However, at atmospheric or elevated temperatures up to the transformations increase in tin from 0 to about 15 per cent resulted in an increase in the dynamic hardness.

---

[11] Upthegrove, C., White, A. E., "A symposium on the effect of temperature upon the properties of metals," Proc. Am. Soc. Test. Matls., **24**, p. 9; 1924; Proc. Am. Soc. Mech. Engrs., **46**, p. 349; 1924.

[12] Bregowsky, I. M., Spring, L. W., "The effect of high temperature on the physical properties of some alloys," Proc. Int. Assoc. Test. Matls., 6th Congress, **2**, VII$_1$; 1912.

[13] Huntington, A. K., "The effect of temperatures higher than atmospheric on tensile tests of copper and its alloys and a comparison with wrought iron and steel," J. Inst. Met., **8**, p. 126; 1912.

[14] Rudeloff, M., "Untersuchungen uber den Einfluss der Wärme auf die Festigkeitseigenschaften von Metallen," Mitteilungen aus den kgl. Versuchsanstalten zu Berlin, **11**, p. 292; 1893.

[15] Rudeloff, M., "The influence of increased temperature on the mechanical qualities of metals," Proc. Int. Assoc. Test. Matls., 5th Congress, **1**, VI$_1$; 1909.

[16] Matsuda, T., "On the dynamic hardness of bronze, aluminum bronze, and brass at high temperature," Sci. Rpts. Tohoku Imp. Univ., **13**, No. 4, p. 401; 1925.

An idea may be obtained of the effect of lead on the high temperature mechanical properties of bronzes from comments of Upthegrove and White [17] relating to copper-tin-zinc-lead alloys of the "steam bronze" type:

> Alloys of the leaded type invariably show loss in tensile strength and decrease in ductility when the melting point of the lead is approached, as indicated by the decrease in strength and ductility slightly below 600° F. (315° C.).

IV. ALLOYS TESTED AND THEIR PREPARATION

Two groups of copper-tin-lead alloys were studied. Their chemical compositions are given in Table 1. In the first group the copper-tin ratio was approximately 92.5 to 7.5 and the proportion of lead was varied from 0.25 to about 25 per cent. In the second group the copper-lead ratio was approximately 84 to 16, and the proportion of tin was varied from a trace to about 10 per cent.

TABLE 1.—*Chemical compositions of the bearing metals tested (per cent)*

[NOTE.—S, denotes sand-cast alloys; C, chill-cast alloys; < less than; n. d., not detected]

Lot No.	Cu	Sn	Pb	Sb	Ni	Fe	Zn	Total	Used for tests indicated
9S	93.2	6.4	0.25	0.06	<0.005	<0.03	n. d.	99.94	Wear and mechanical properties.
9C	93.1	6.5	.25	.06	<.005	<.03	n. d.	99.94	
1XS	88.2	8.3	3.4	.03	<.005	<.03	n. d.	99.96	Wear.
1XC	88.2	8.3	3.4	.03	<.005	<.03	n. d.	99.96	
10S	87.6	7.4	4.7	.14	.05	<.03	n. d.	99.91	Wear and mechanical properties.
10C	87.5	7.4	4.8	.13	.05	<.03	n. d.	99.90	
11S	83.8	6.3	9.6	.17	.07	<.03	n. d.	99.96	Wear.
11C	83.7	6.3	9.7	.17	.07	<.03	n. d.	99.96	
12S	78.8	6.3	14.6	.15	.06	<.03	n. d.	99.93	Wear and mechanical properties.
12C	78.8	6.2	14.7	.16	.06	<.03	n. d.	99.94	
13S	73.3	5.3	20.9	.27	.11	<.03	n. d.	99.90	Wear.
13C	73.3	5.3	20.9	.29	.12	<.03	n. d.	99.93	
14S	70.4	5.45	23.9	.13	.04	<.03	n. d.	99.94	Do.
14C	70.2	5.3	24.3	.12	.04	<.03	n. d.	99.98	
22S	70.3	5.1	24.5	.03	<.005	<.03	n. d.	99.96	Mechanical properties.
22C	70.3	4.9	24.7	.03	<.005	<.03	n. d.	99.96	
15S	84.4	.73	14.7	.02	.08	<.03	n. d.	99.95	Wear.
15C	84.9	.68	14.3	.02	.08	<.03	n. d.	100.00	
23S	86.1	.02	13.8	.01	<.005	<.03	n. d.	99.96	Mechanical properties.
23C	84.6	.30	14.9	.07	n. d.	<.03	n. d.	99.89	
21S	83.7	2.4	13.8	.02	n. d.	<.03	n. d.	99.94	Wear and mechanical properties.
21C	84.2	2.35	13.4	.02	n. d.	<.03	n. d.	99.99	
16S	80.1	5.7	13.8	.24	0.08	<.03	n. d.	99.94	Wear.
16C	80.2	5.65	13.8	.23	.08	<.03	n. d.	99.98	
24S	83.5	5.2	11.2	.03	n. d.	<.03	n. d.	99.95	Mechanical properties.
24C	80.8	4.3	14.7	.13	n. d.	<.03	n. d.	99.95	
17S	78.9	7.9	12.9	.17	0.04	<.03	n. d.	99.93	Wear.
17C	78.9	7.8	13.0	.17	.04	<.03	n. d.	99.93	
18S	77.6	10.1	12.0	.18	.03	<.03	n. d.	99.93	Do.
18C	77.6	10.0	12.1	.18	.03	<.03	n. d.	99.93	
25S	75.5	9.8	14.5	.03	<.005	<.03	0.13	99.99	Mechanical properties.
25C	75.4	9.8	14.5	.03	<.005	<.03	.15	99.91	

The alloys were cast into "sticks" about 12 inches long and 2½ inches in diameter. Originally, commercial mixtures of journal bearings, scrap engine castings, railroad turnings, ingot metal, and

[17] See footnote 11, p. 352.

some copper, tin, or lead were melted in a Fisher furnace and the resulting metal poured by the molder who had made the molds.

Both the chill-cast and the sand-cast "sticks" were cast "on end" and were gated at the bottom. Details of the chill mold are shown in Figure 4, where dotted areas represent green sand and cross-hatched areas represent cores. Similar equipment was used for the sand castings, the split chill mold being replaced by a sand mold. The pouring temperature was kept around 2,050° F. (1,120° C.) for the sand castings and around 1,900° F. (1,035° C.) for bronzes cast in the metal molds, which were preheated to 350° F. (175° C.).

When machined into test specimens, the "sticks" prepared in the manner described were found to be filled with tiny gas pockets and

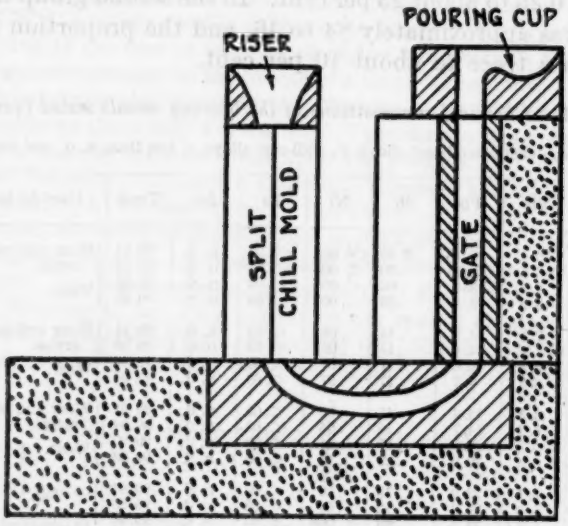

FIG. 4.—*Chill-mold used*

The thickness of mold wall was 1 inch.

were therefore discarded. Different methods of melting and degasification were next tried, including additions of magnesium, calcium carbide, manganese dioxide or barium sulphate, but the resultant castings were not entirely free from similar imperfections.

The method next tried, and one which gave castings practically free from gas pockets and porosity, was as follows: Ingots of two compositions were prepared from ingot copper, pig tin, and pig lead. One group contained about 92.5 per cent copper and 7.5 per cent tin, and a second group contained about 84 per cent copper and 16 per cent lead. Ingots of the first group, with pig lead to make a desired composition, or ingots from the second group, with the required pig tin, were melted in crucible furnaces without covering. The temperature was brought to about 2,100° F. (1,150° C.) and the fuel supply

shut off, the melt then being allowed to freeze with only occasional stirring. When the surface of the melt could no longer be penetrated by the stirring rod, it was covered with glass and again melted. The metal was then poured at a temperature around 2,050° F. (1,120° C.).

While this method of melting, freezing, and then remelting under glass gave satisfactory results, it was considered too slow for the production of any appreciable amount of test material. Most of the alloys were melted under glass in a Fisher crucible furnace under regular shop conditions. The ingots prepared in the manner already described were used as the basis of the melt. The sand-cast "sticks" were poured "on end" through the top, the mold being rammed up entirely in the drag half of the flask and not being covered at all. The chill castings were made in a similar manner in the metal molds, with the exception that a cup was placed on top of the mold.

It is seen from the foregoing that several methods were used in the production of the test castings at the manufacturer's plant. The manufacturer did not keep a record of the alloys produced by each method, and therefore such information could not be included in this report. However, close inspection was made of the castings to guarantee freedom from minute gas holes which might not be detected in commercial inspection. This was considered of greater importance than a record of whether individual castings were poured through horn gates or top poured.

Variations in structure and properties are to be expected from the outside to the center of the castings so that all specimens for chemical analysis,[18] metallographic examination, and the mechanical tests were taken from areas representative of the metal at the surfaces of the wear-test specimens, as is shown in detail in Figure 5. Thus, comparable conditions were studied.

The microstructures of the different alloys are shown in Figures 6 to 11, inclusive. Figures 6 and 7 represent the unetched specimens, and show principally the size, amount, and distribution of the lead particles. The number of these increases with the proportion of lead, and they are generally smaller in size in alloys cast in metal molds than in the corresponding alloys cast in sand molds.

Figures 8 to 11, inclusive, show the structures of etched specimens. The alloys consist of a matrix of alpha grains plus varying amounts of the alpha-delta eutectoid, depending upon the proportion of tin present and the method of casting. The insoluble lead particles are likewise distributed throughout the matrix.

According to the customary copper-tin equilibrium diagrams, about 12 per cent tin is required for the first appearance of the eutectoid.

[18] Made by H. A. Buchheit, assistant chemist.

This is equivalent to a tin-copper ratio of approximately 0.14.[18] Under the casting conditions encountered in the preparation of the test ingots, which do not represent equilibrium conditions, the eutectoid was found in alloys containing around 7.5 to 8 per cent tin, which is equivalent to a tin-copper ratio in the matrix of about 0.09+. A greater proportion of this eutectoid was generally found in the alloys cast in metal molds than in the corresponding alloys cast in sand

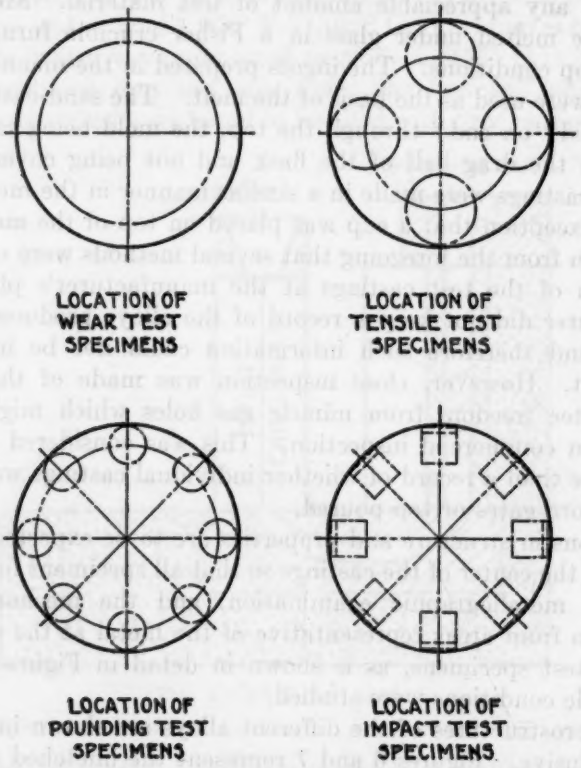
LOCATION OF WEAR TEST SPECIMENS
LOCATION OF TENSILE TEST SPECIMENS
LOCATION OF POUNDING TEST SPECIMENS
LOCATION OF IMPACT TEST SPECIMENS

FIG. 5.—*Locations of the test specimens with respect to the test castings*

molds, as will be evident from examination of Figures 8 and 9. However, the eutectoid areas present in the sand castings were usually the larger. There was likewise an increase in the amount of the eutectoid as the tin was raised from around 7.5 to 10 per cent.

Thus, the alloys containing less than around 7.5 per cent tin consisted of alpha grains with particles of lead distributed throughout the matrix and no readily visible areas of the eutectoid; alloys con-

[18] This ratio is still in dispute. Stockdale's recent studies reported in the 1925 J. Inst. of Metals, **34**, p. 111, show the range of alpha-solid solubility to be 16 per cent tin with long-time annealing. However, this does not alter the fact that the eutectoid is found under commercial casting conditions with much lower tin concentration than that required under equilibrium conditions.

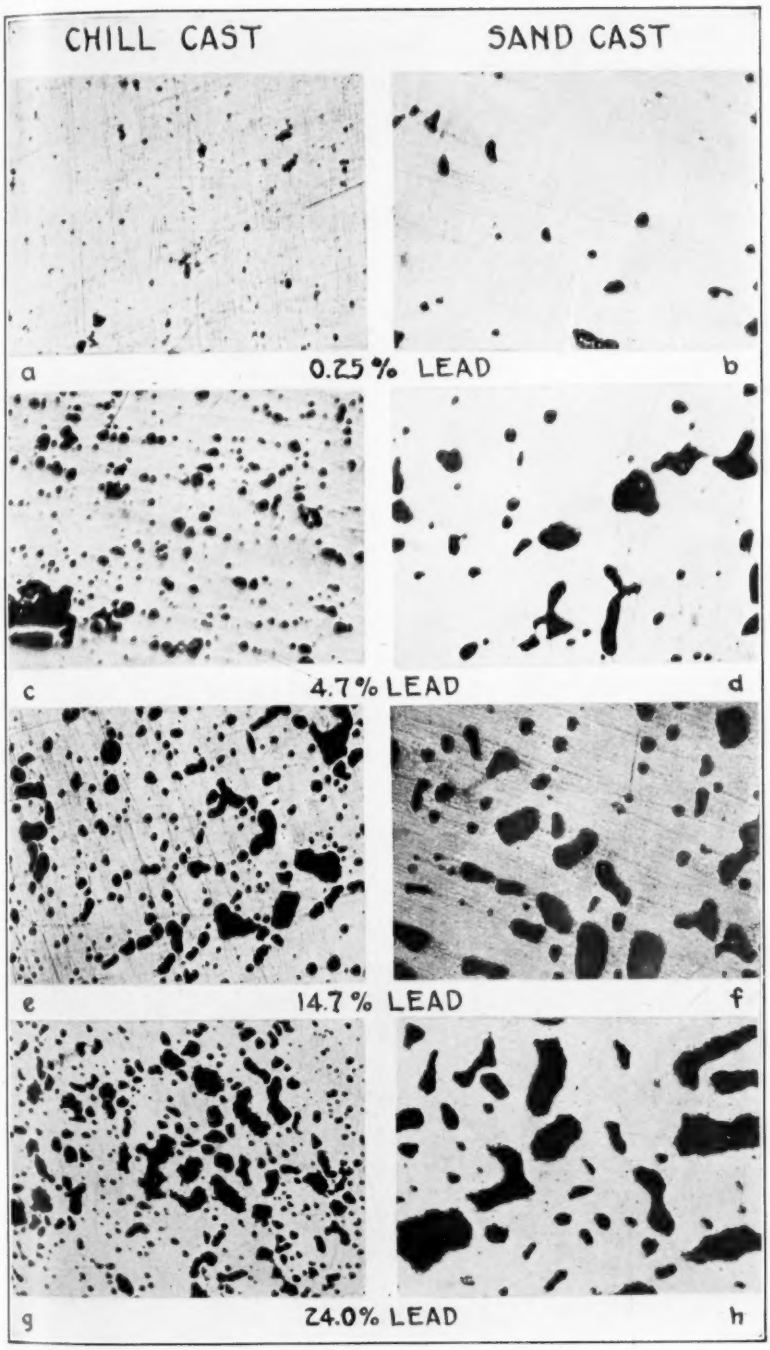

FIG. 6.—*Photomicrographs of unetched specimens of the bronzes varying in lead.* ×100

UNIV. OF MICH.

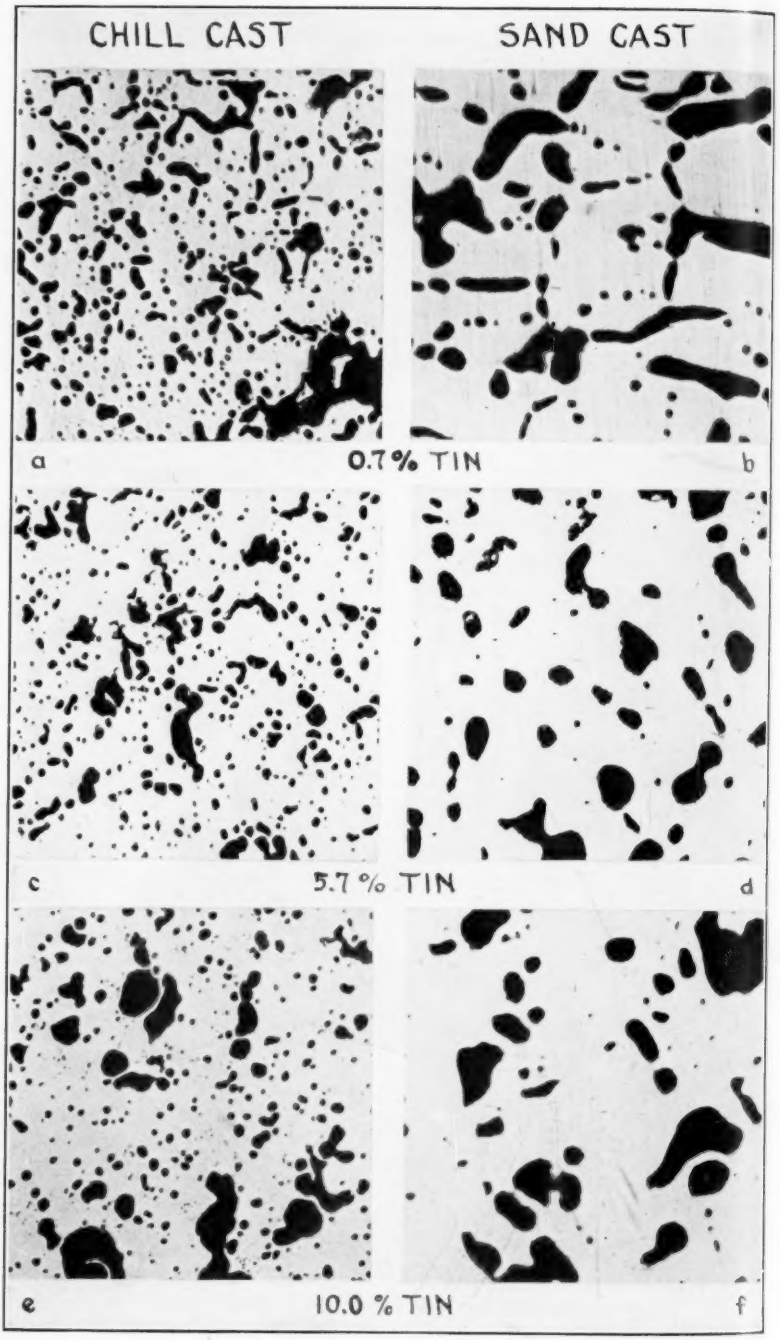

FIG. 7.—*Photomicrographs of unetched specimens of the bronzes varying in tin.* ×100

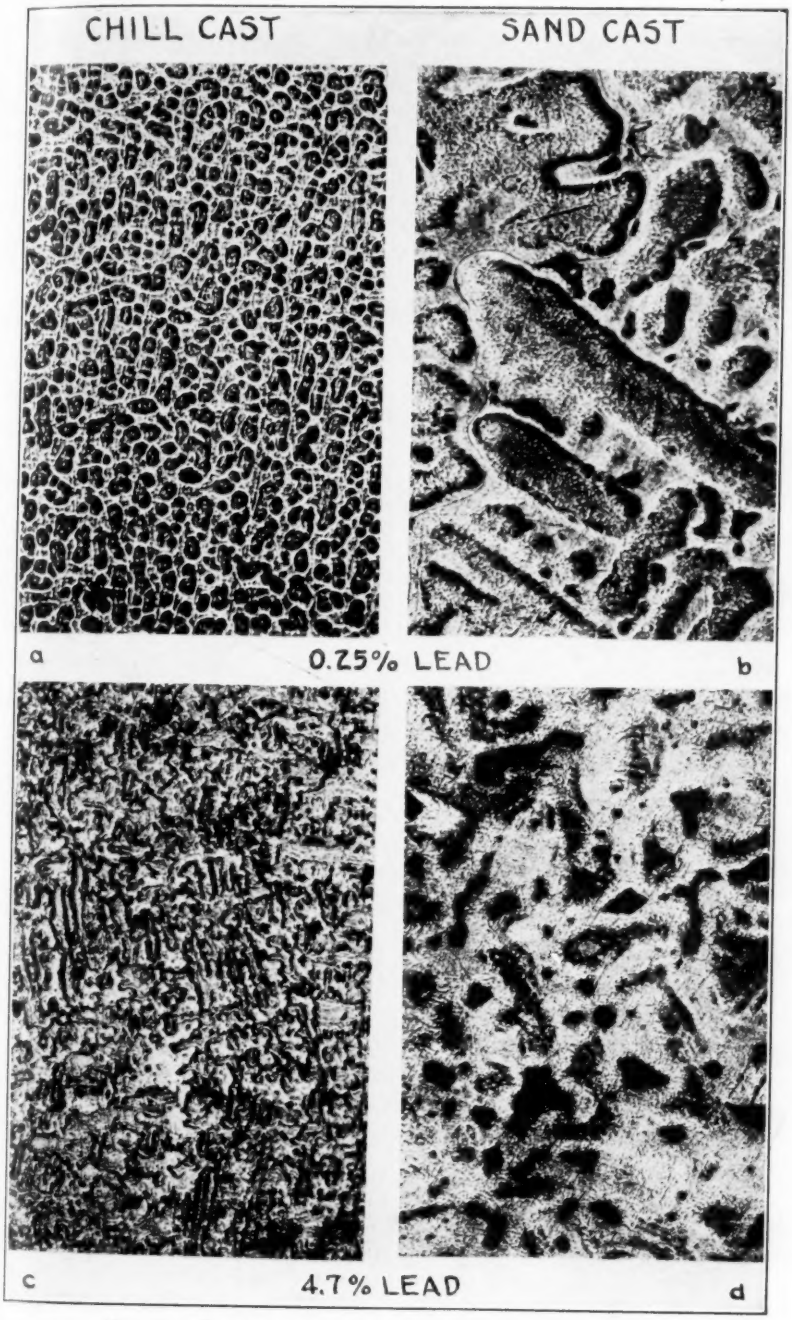

FIG. 8.—*Structure of some of the bronzes varying in lead*

Etched with solution of 3 parts NH_4OH plus 1 part H_2O_2. Then etched with acidic ferric chloride solution. ×100

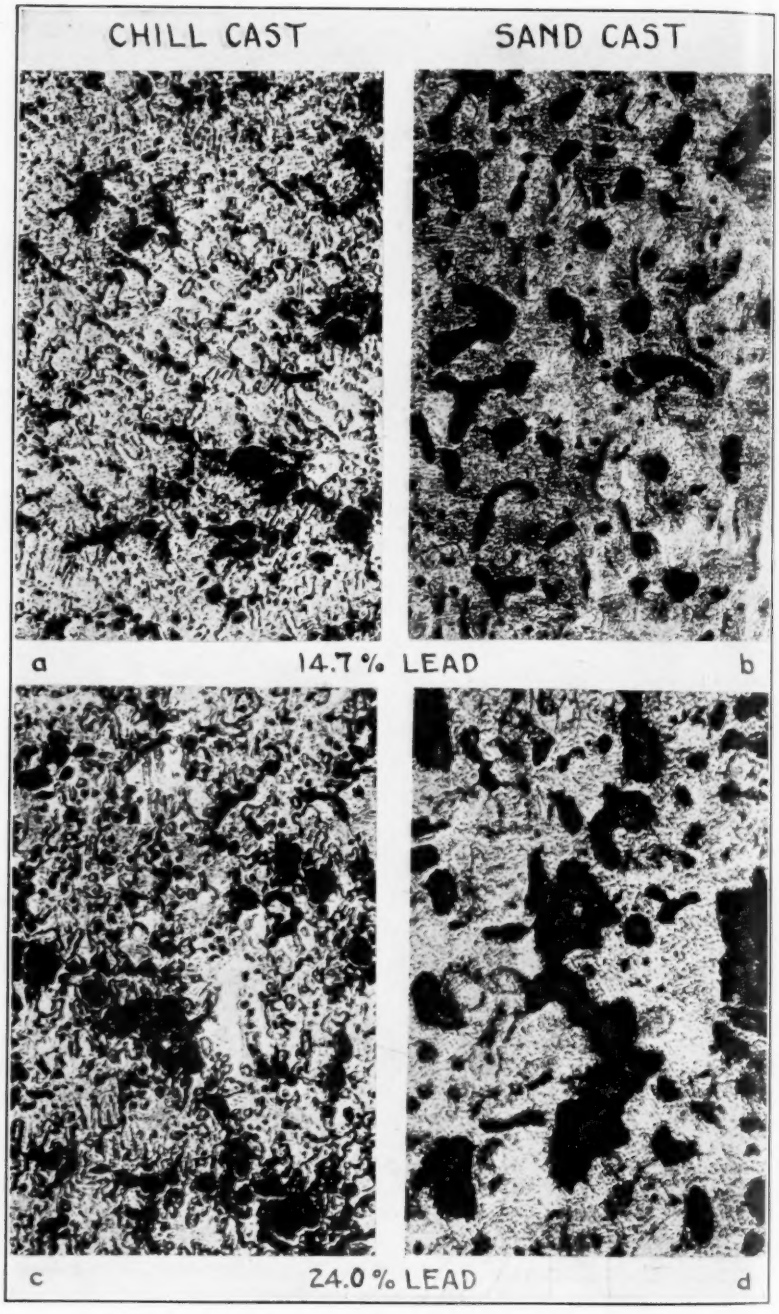

FIG. 9.—*Structure of some of the bronzes varying in lead*

Etched with solution of 3 parts NH_4OH plus 1 part H_2O_2. Then etched with acidic ferric chloride solution. ×100

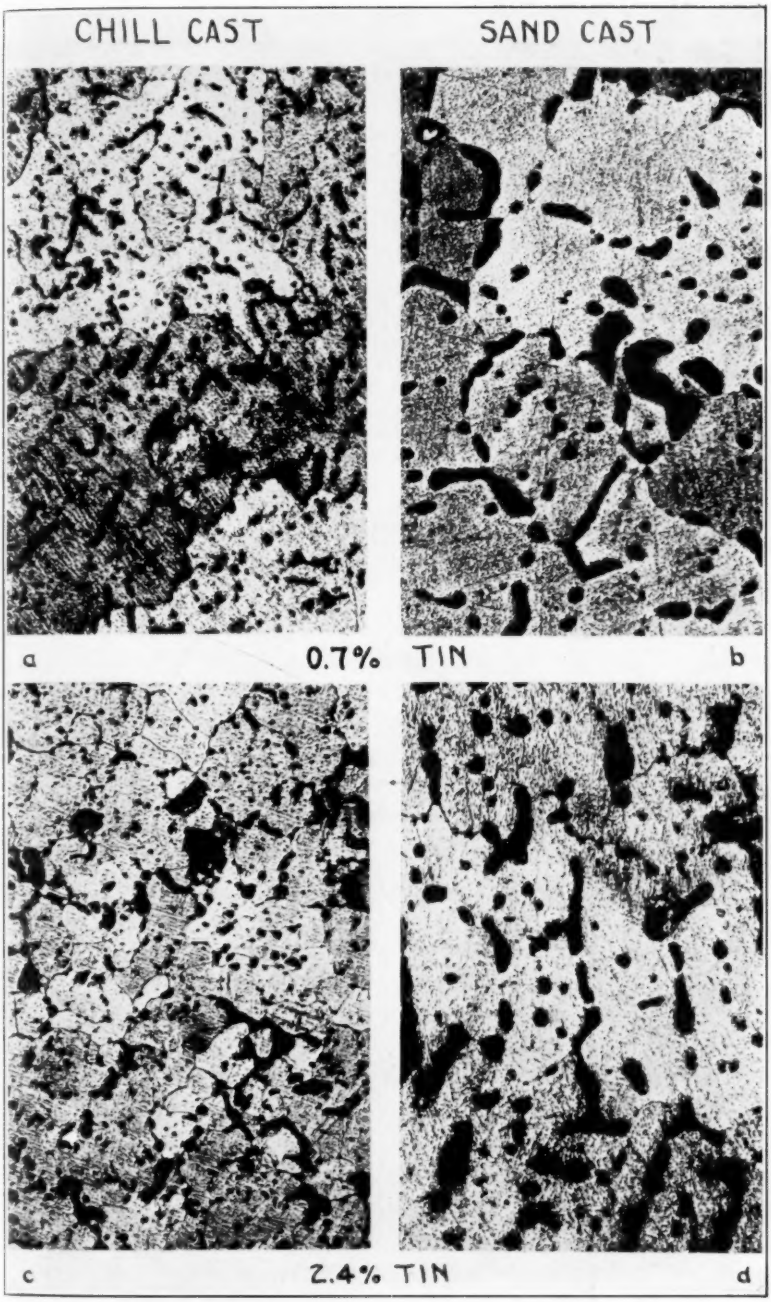

FIG. 10.—*Structure of some of the bronzes varying in tin*

Etched with solution of 3 parts NH_4OH plus 1 part H_2O_2. Then etched with acidic ferric chloride solution. ×100

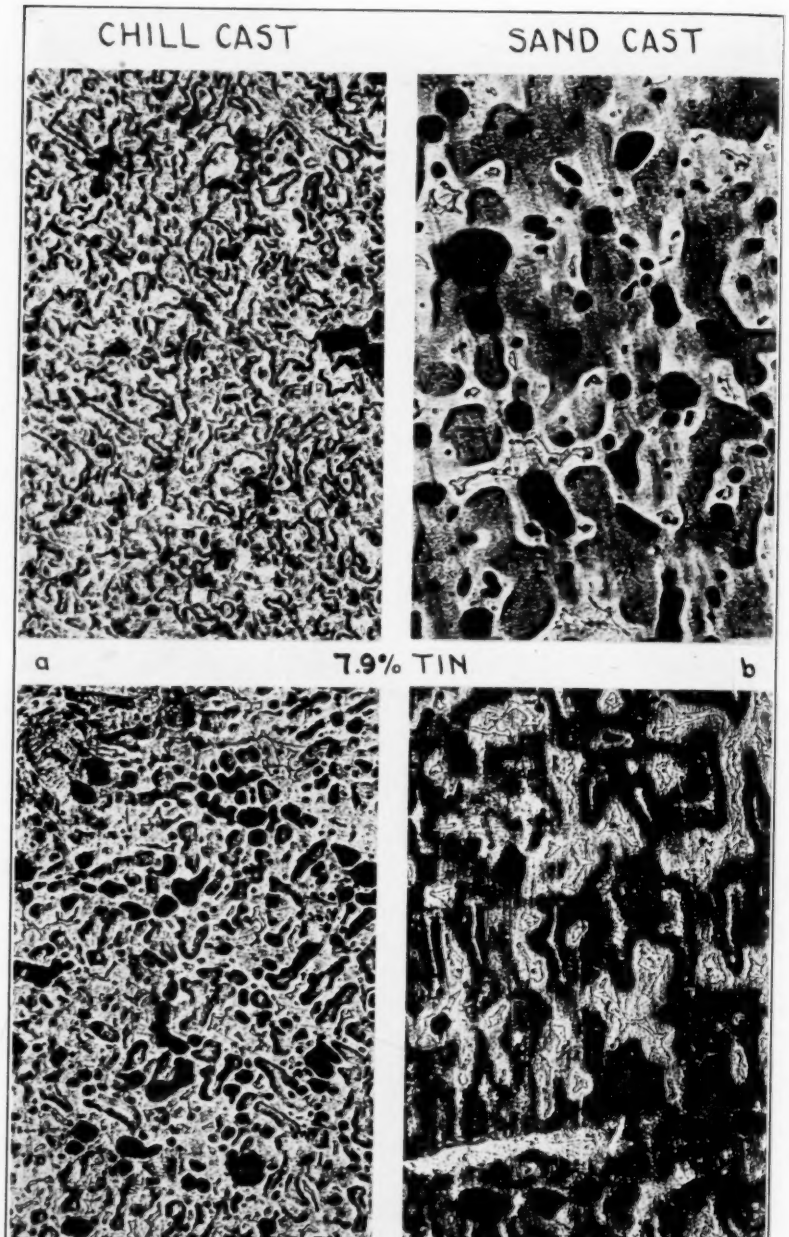

FIG. 11.—*Structure of some of the bronzes varying in tin*

Etched with solution of 3 parts NH_4OH plus 1 part H_2O_2. Then etched with acidic ferric chloride solution. ×100

taining around 7.5 per cent or more of tin consisted of alpha grains, the eutectoid, and particles of lead.

The steel specimens against which the bronzes were rotated in the wear tests were taken from the center of used railroad car axles. The first axle selected, which will be designated as steel A, was not large enough to give the required number of specimens, and so the last few groups of wear tests were made with steel specimens taken from a second axle, designated as steel B. This had a chemical composition comparable to that of the axle originally used.

Details of the chemical compositions and some of the properties of the steels used in the wear tests are summarized in Table 2. Since the specimens were cut from car axles, they represent the steels with which bearings are in contact in railroad service.

TABLE 2.—*Chemical composition and other details of the steels used in contact with the bronzes in the wear tests*

Composition and properties	Steel A	Steel B
	Per cent	*Per cent*
Carbon	0.37	0.44
Manganese	.51	.42
Phosphorus	.025	.008
Sulphur	.026	.040
Silicon	.02	.10
Chromium	n. d.	n. d.
Vanadium	n. d.	n. d.
Nickel	n. d.	n. d.

n. d. = not detected in chemical analysis.

Source: Steels A and B taken from used railroad car axles.
Microstructure: Steel A, moderately large grains of pearlite in ferrite matrix; steel B, same as A, but larger grain size.
Brinell hardness number: Steel A, 153; steel B, 129.

V. EQUIPMENT AND TEST METHODS EMPLOYED

1. THE AMSLER WEAR TESTING MACHINE

The machine selected for the study of the wear of railroad bearing bronzes in this investigation is the one manufactured by Alfred J. Amsler, Schaffhouse, Switzerland, and is shown in Figure 12. Heretofore, it has been used largely in studying the wear of ferrous metals, but Pecoraro [20] has reported its application to babbitts.

The essential features of the Amsler wear testing machine are (1) a gear train so constructed as to give either sliding or rolling friction, (2) a calibrated spring P (fig. 12) for adjusting the contact pressures between the test specimens, (3) a cam device C (fig. 12) which may be used to give a lateral sliding motion between the test specimens during test, and (4) a friction dynamometer and torque indicator T (fig. 12) for determining the work done upon the test specimens and the torque developed during the test.

[20] Pecoraro, N., "Experimental tests of white metals for engines," Intl. Assoc. Test. Matls., 6th Congress, II$_{2c}$; 1912.

The motor driving the gear train may be operated at two speeds, depending upon whether it is connected to 110 or 220 volt lines. Only the low speed was used in the tests to be described in this report.

For the wear tests made under combined rolling and sliding friction the specimens used were in the form of disks initially 2 inches in diameter and 0.4 inch thick. These were mounted, respectively, one above the other on the two spindles S_1 and S_2 shown in Figure 12 and revolved in opposite directions (clockwise and counterclockwise), so that at the line of contact the surfaces of the two disks were moving in the same direction but at different speeds. There was, therefore, a positive slip between the disks. Under the conditions of operation the lower specimen S_2 had an angular velocity of approximately 240 r. p. m. and the speed of the upper specimen S_1 was equal

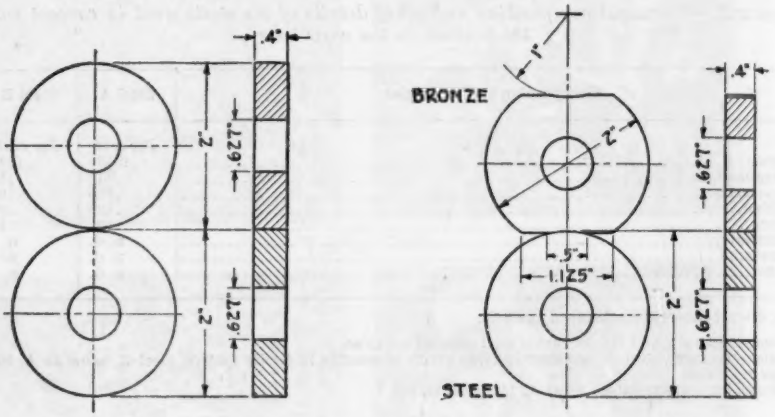

FIG. 13.—*Form and dimensions of the wear-test specimens used*

to 0.905 times the speed of the lower one. Thus, the surface speed of the lower specimen was approximately 126 feet per minute and the slip between the two specimens 12 feet per minute. In all tests the lateral oscillating motion produced by the cam was used. The amplitude of oscillation of the upper specimen S_1 was about 0.3 inch, and there were about four oscillations for each three revolutions of the upper test specimen.

For the wear tests made under sliding friction the gear actuating the upper shaft was disengaged, and the shaft was locked in place. The upper specimen used for such tests had a concave contact surface machined to conform to the surface of the lower specimen. The form and dimensions of the test specimens for these tests are shown in Figure 13.

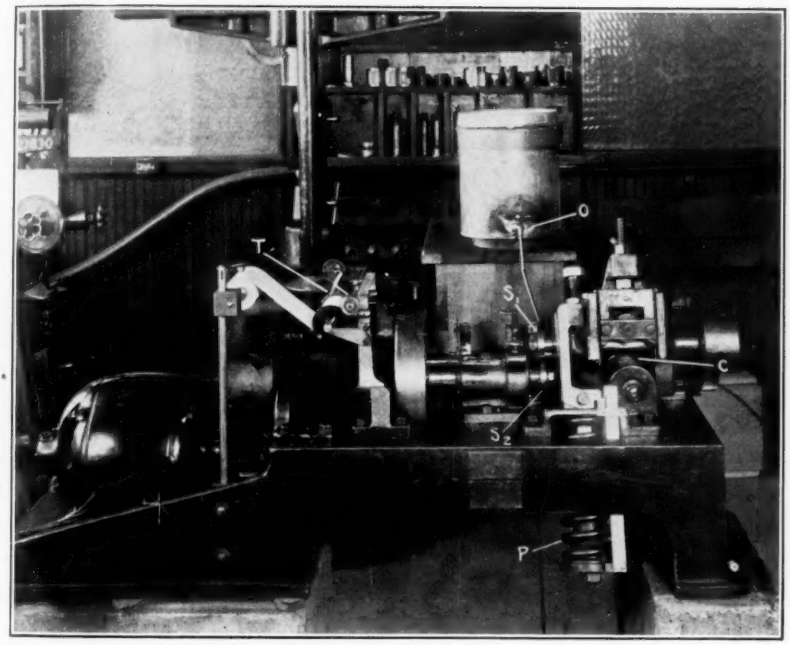

FIG. 12.—*The Amsler wear testing machine used in the experiments*

S_1 and S_2, Test specimens
P, Calibrated spring for controlling contact pressures
C, Cam for producing lateral oscillation of test specimen S_1
E, Torque indicator
O, Oil supply reservoir for tests under lubrication

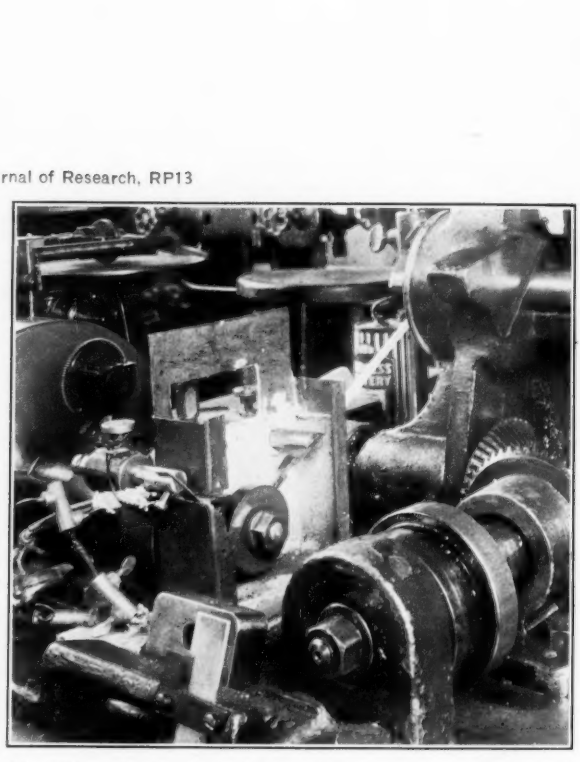

FIG. 14.—*Furnace used in making wear tests at elevated temperatures*

This furnace is constructed in two sections. The upper half slides on the lower half which is shown in the illustration. Note the provision made for using the copper scrapers

The bronze specimen shown in Figure 13 for use under sliding friction is only satisfactory in cases where the total wear is low, as in the wear tests with lubrication which were made in this investigation. When appreciable wear takes place, the contact area becomes larger and progressively changes the contact pressures. These difficulties can be avoided by using specimens with straight sides, so that the contact area will remain the same whether large or small weight losses are encountered.

The range of contact pressures which could be exerted between the specimens was from 16.7 to 300 kg (37.5 to 660 pounds). The lowest pressure was obtained from the weight of the head carrying the upper spindle of the testing machine, while higher pressures could be obtained by means of the calibrated spring *P*. (Fig. 12.) All tests reported were made under contact pressures produced by the weight of the head carrying the upper spindle, namely, 16.7 kg (37.5 pounds) total load.

The Amsler wear testing machine is also provided with an auxiliary cam which when engaged periodically releases the load by raising the head carrying the upper spindle and then lets it drop. This produces a repeated pounding of the test specimens. However, wear tests were not made under this pounding action in this investigation. The resistance of the different metals to deformation was studied in mechanical tests.

By changing weights on the pendulum dynamometer, three ranges of sensitivity can be secured, namely 50, 100, and 150 kg-cm (43.4, 86.7, and 130 lb.-in.). An integrating counter turns on a disk directly connected to the gear train of the testing machine. This counter is connected to an arm attached to the pendulum, and its position on the disk is dependent upon the position of the pendulum which indicates the frictional forces between the specimens. Thus, the reading of the integrating counter is a measure of the work done upon the specimens.

A continuous torque record is obtained on a revolving drum by means of a pencil connected to the pendulum.

For tests with lubricants a chain is provided which can be suspended from one of the spindles in a pan of the lubricant. Friction between the spindle and the chain puts the latter in motion, and the chain then carries the lubricant from the reservoir to the spindle and the specimen. However, this equipment was replaced in all of the wear tests made in the presence of lubricants in this investigation. Oil was supplied drop by drop to the upper specimen through a small tube connected to an oil supply reservoir *O*. (Fig. 12.) The rate of flow was kept at about 60 drops per minute by means of a suitable valve. In this manner, a continuous supply of fresh oil was fed to the specimens, and the used oil frequently containing abraded metal

particles was collected in a pan at the base of the machine and not used again.

In tests at 175° and 350° F. (80° and 175° C.) the specimens were subjected to combined rolling and sliding friction within a small electric resistance furnace. A photograph of this furnace is reproduced in Figure 14. The resistances were placed at both the top and bottom of the heating chamber, and a Pyrex glass window was provided through which the specimens might be observed during the progress of the tests. Openings in the side walls permitted the use of the copper scrapers as in the tests at atmospheric temperatures. A thermometer and likewise a thermocouple and potentiometer were used in determining temperatures.

The spindle carrying the upper specimen in the testing machine was raised slightly, so that while being heated to the desired temperatures the test specimens could be rotated out of contact. This minimized temperature changes encountered in first starting rotation of the disks when the furnace had been brought to the desired temperature with the specimens at rest. After about 40 minutes at temperature the support was removed from the upper spindle and the rotation of the bronze and steel in contact was begun. From this point the procedure was similar to that used in the wear tests at atmospheric temperatures.

There is not much clearance between the spindle bearings of the Amsler wear testing machine for the installation of a heating unit, and difficulties were encountered in the first high temperature wear tests. The furnace illustrated in Figure 14 was mounted directly on the faceplate of the housing of the lower spindle bearing. Conduction of heat increased the temperature of the bearing and brought oil vapor from the bearing into the heating chamber of the furnace. This oil was deposited on the test specimens and through its lubricating action materially affected the rates of wear.

These difficulties were minimized by modifications in some of the details of construction of the bearing, namely, (1) a felt washer which quickly became soaked with oil was replaced by an aluminum washer having a very small clearance with the shaft, (2) the aluminum faceplate having a relatively high thermal conductivity was replaced by one made of bakelite, (3) a hole was drilled in the new faceplate from the bottom to the spindle and connected to a suction line through which oil drops or oil vapors were removed before they could enter the heating chamber and lubricate the specimens.

It was impracticable to prove that all oil was kept away from the test specimens, but the described changes resulted in increased wear of the bronzes, no oil was detected on the specimens, and it was possible to reproduce results. The major difficulties were, therefore, considered to have been removed.

(*a*) Calculation of Unit Contact Pressures.—The chief difficulty in calculating the unit contact pressures for the two cylindrical specimens used in the wear tests under combined rolling and sliding friction is in the determination of the contact areas. The thickness of the specimens definitely determines the length of the contact area, but the width of contact is a function of the radii of the test specimens, the physical properties of the metals under test, and the total pressure between the specimens.

The equations of H. Hertz [21] for the determination of the width of contact and of the unit stress existing between two curved surfaces under load were used to calculate the unit contact pressures. These equations are as follows:

$$b=4\sqrt{\frac{P}{\pi l}\times\frac{\frac{1}{E_1}(1-m_1^2)+\frac{1}{E_2}(1-m_2^2)}{\frac{1}{R_1}+\frac{1}{R_2}}} \tag{1}$$

and

$$S=\frac{4P}{\pi bl} \tag{2}$$

where E_1 and E_2 are the moduli of elasticity of the metals from which the specimens are made, m_1 and m_2 represent Poisson's ratios for these metals. R_1 and R_2 are the radii of the two specimens, l is the length of the area of contact between the two specimens, b is the width of the area of contact between the two specimens, P is the total pressure between the two specimens, and S is the unit stress existing between the specimens.

These equations are based on a number of assumptions, the most important of which are: (1) The contact area is small compared with the dimensions of the cylinders. (2) There are no shearing stresses at the contact surfaces; that is, only stresses normal to the contact surfaces are considered. (3) The stresses are within the limit of elasticity of the metals.

By substituting in equation (2) the value of b determined in equation (1), equation (3) is obtained:

$$S=\frac{P}{\pi l}\times\frac{1}{\sqrt{\frac{\frac{P}{\pi l}\left[\frac{1}{E_1}(1-m_1^2)+\frac{1}{E_2}(1-m_2^2)\right]}{\frac{1}{R_1}+\frac{1}{R_2}}}} \tag{3}$$

[21] H. Hertz, Misc. Papers. MacMillan & Co. (Ltd.), p. 163; 1896. A more convenient reference is F. H. Buhlman, "Factors governing the design of roller bearings for roll necks," Iron and Steel Engr., p. 3045; June 1927.

For specimens of given dimensions and materials, equation (3) may be reduced to

$$S = C\sqrt{P} \tag{4}$$

in which C is a constant which may be calculated from

$$C = \sqrt{\frac{\frac{1}{R_1}+\frac{1}{R_2}}{\pi l\left[\frac{1}{E_1}(1-m_1^2)+\frac{1}{E_2}(1-m_2^2)\right]}} \tag{5}$$

The unit stresses existing between the bronze and steel specimens tested were determined by the above equations, using the following values for the constants:

$E_1 = 29{,}500{,}000$ lbs./in.^{2} (steel).
$E_2 = 12{,}500{,}000$ lbs./in.^{2} (bronzes).
$m_1 = 0.3$ (steel).
$m_2 = 0.33$ (bronze).
$R_1 = R_2 = 1$ inch.
$l = 0.4$ inch (maximum length of contact area).
$l = 0.25$ inch (minimum length of contact area when using the lateral oscillating motion previously referred to).

The total pressure between the specimens during the tests was about 37.5 pounds. The minimum unit contact pressure between the specimens, when l was 0.4 inch, was 24,200 lbs./in.^{2}, and the maximum unit contact pressure, during the lateral oscillation, when l was 0.25 inch, was 30,600 lbs./in.^{2}

It should be noted that these stresses are only an approximate determination of the true stresses due to the fact that various assumptions have been made, as previously outlined, and that the physical constants of the metals are not accurately known. It should also be pointed out that the proportional limits of the bronzes tested were less than the computed stresses. Therefore, plastic deformation will take place, and the actual unit contact pressures should be less than the computed values. However, the calculations made give a useful, though qualitative, characterization of some of the conditions of test.

2. SAND ABRASION TESTS

When placed in railroad service, bearing bronzes are frequently subjected to the abrasive action of nonmetallic particles—as, for example, road dust. It was therefore considered advisable to determine whether some of the major differences observed in the lubricated and unlubricated wear tests on the Amsler machine would be maintained when wear was produced by a nonmetallic abrasive. A sand blast, designed for sand abrasion tests, was used for these

determinations. Abrasion is produced in this test by the blasting of sand upon the specimen instead of rubbing the nonmetallic particles upon the lubricated bronze, as actually occurs in service, but it was thought that some useful information could be obtained.

In this test a special nozzle, five-sixteenths inch diameter, lined with rubber to minimize abrasion of the nozzle, was placed in the top of a wooden box, the interior of which was also lined with rubber. The nozzle was connected to the bottom of a Y pipe fitting. Sand and air were introduced through the two upper branches of the Y, the air passing through an inner tube in the Y. The lower end of this tube, which carried the air blast into the nozzle, was below the point at which the sand was introduced so as to avoid "backing up" of the sand. The introduction of the sand was controlled by a slide valve.

The Amsler wear test specimens shown at the left side of Figure 13 were used for these tests, but the holes were plugged with rubber. During the tests the specimens were mounted in the box shown in Figure 15. The flat faces were placed at an angle of 45° to the axis of the nozzle, the center of the specimens being located about 2½ inches from the nozzle tip. An opening in the bottom of the box provided an outlet for the sand and air.

One kilogram (2.2 pounds) of standard Ottawa silica sand, 20 to 30 mesh, was blasted on the specimens with an air pressure of 60 lbs./in.^{2}, and the loss in weight of the specimens was taken as a measure of the wear. Only new sand was used for each test so as to insure a uniform abrading material. The test procedure was to place the weighed amount of sand in a hopper, adjust the air pressure to 60 lbs./in.,^{2} and then open the slide valve which gave an immediate and full flow until all the sand had passed through the nozzle. This required about one minute. Several consecutive tests were made on each specimen until a definite rate of wear was observed.

3. SINGLE-BLOW IMPACT TESTS

The impact tests were made with an Izod impact testing machine on specimens of the form and dimensions shown in Figure 16. In general, six specimens were tested for each condition of each alloy at each temperature. Suitable furnace equipment was provided for the tests at elevated temperatures. This equipment consisted of a preheating furnace, an equalizing furnace (*A*, fig. 17), and a special vise *B*, within which resistance wires were embedded. These wires (shown at *C* in fig. 17) permitted independent control of the temperature of the vise, which could be brought to the same temperature as the test specimen. This tended to reduce the surface chilling encountered in placing heated metal specimens in the jaws of a vise at atmospheric temperatures.

A group of specimens were heated to the desired temperature in the preheating furnace. One specimen was then removed and placed in the heated vise of the impact testing machine. The equalizing furnace (*A*, fig. 17), previously heated to the same temperature, was placed over the specimen and left in this position for about 10 minutes,

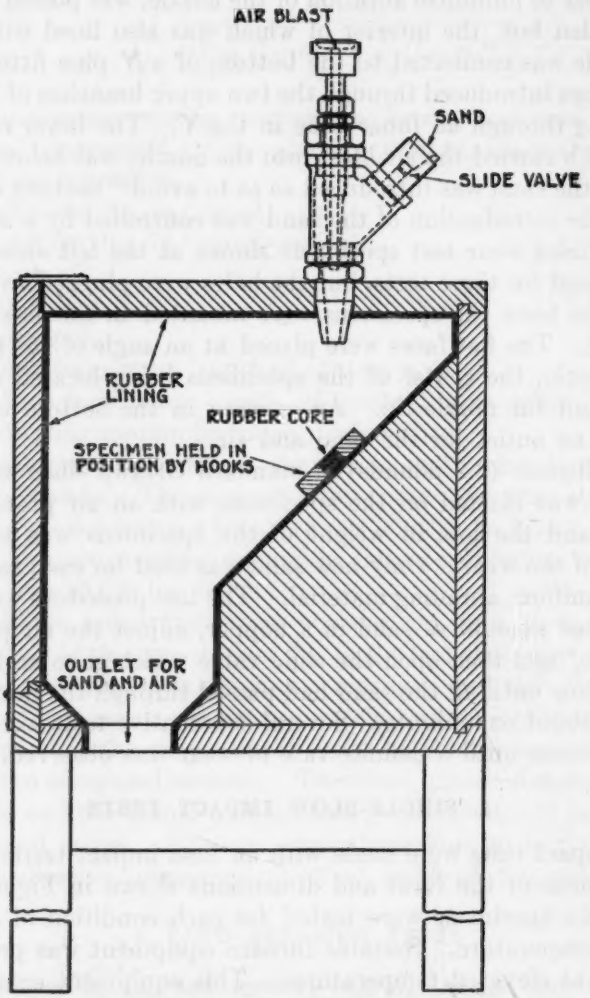

FIG. 15.—*Sand-blast abrasion apparatus*

during which time the surface of the specimen previously cooled by momentary contact with the air was again brought to the desired temperature. The tup *D* was then placed in position for the blow, the equalizing furnace *A* removed, and the tup immediately released. In this way the heated specimen was exposed to the air before test for a period of less than one second. At the same time other specimens

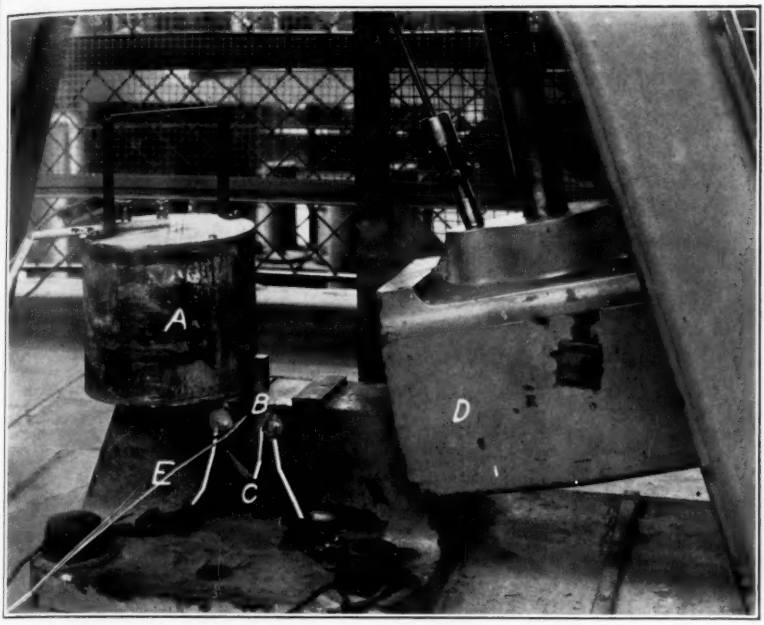

FIG. 17.—*Part of equipment used in the notched-bar impact tests*

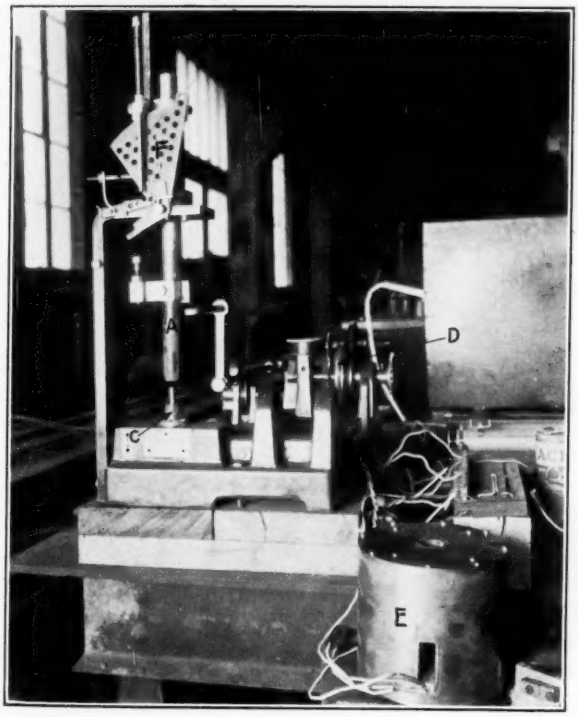

FIG. 18.—*Equipment used in the pounding tests*

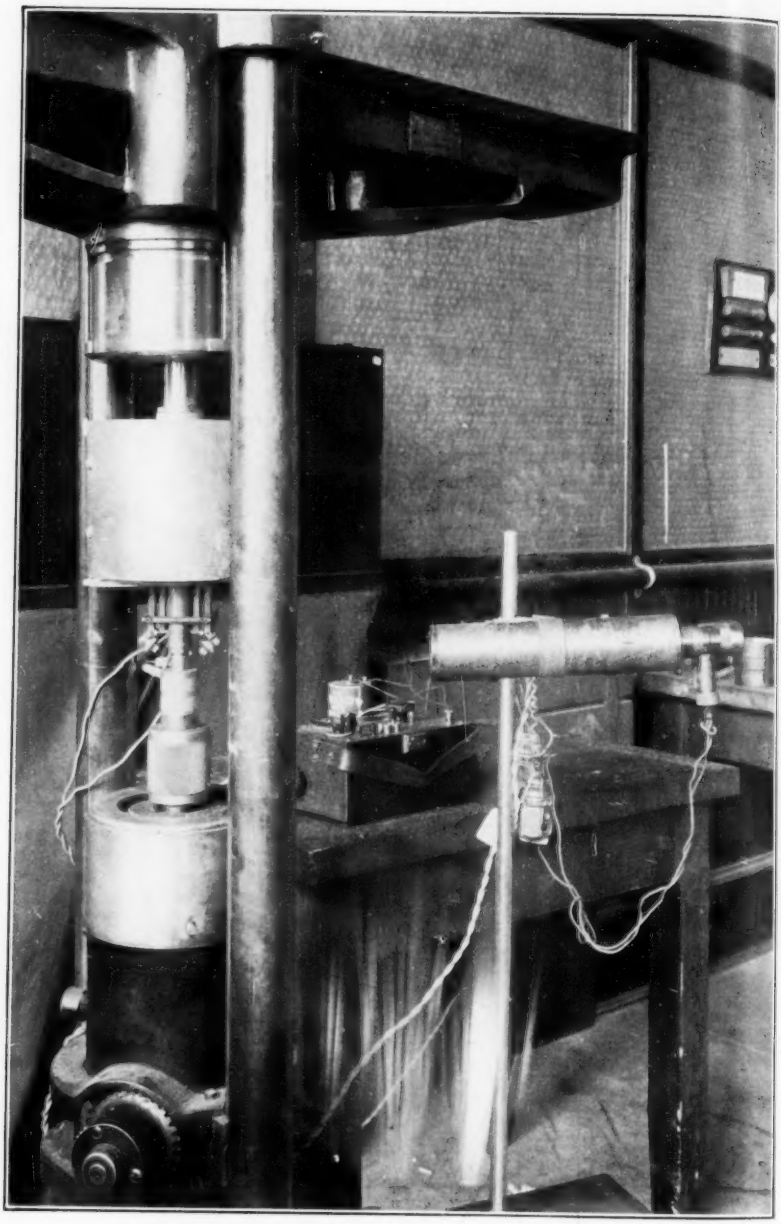

FIG. 19.—*Equipment used in the tension tests*

in the preheating furnace were ready for the short heating for temperature equalization. One of the advantages of this procedure was the large saving in time effected when testing a large number of specimens.

Test temperatures were measured by a potentiometer and thermocouple (*E*, fig. 17) placed in a small hole in the test specimen.

4. TENSION TESTS

The tension tests were made with a hydraulic testing machine and a modified Martens extensometer on specimens of the form and dimensions shown in Figure 16. The equipment, which is illustrated in Figure 19, and the procedure employed in tests at different temperatures have already been described.[22]

5. REPEATED-BLOW POUNDING TESTS

Pounding tests in which small cylindrical specimens were subjected to repeated blows in compression were carried out with the equipment shown in Figure 18. This consisted of a Stanton repeated impact testing machine modified to give repeated blows in compression to a specimen mounted on the anvil. In Figure 18, *A* is the falling weight, *B* is the specimen, *C* is the anvil, *D* the operating mechanism of the Stanton repeated impact testing machine, and *E* the furnace within which the specimens were tested at different temperatures. The mechanism at *F* is to insure individual blows free from the auxiliary blows associated with the rebound of the falling weight. This was not taken care of in the Stanton machine and formed an important part of the modified equipment[23] shown in Figure 18.

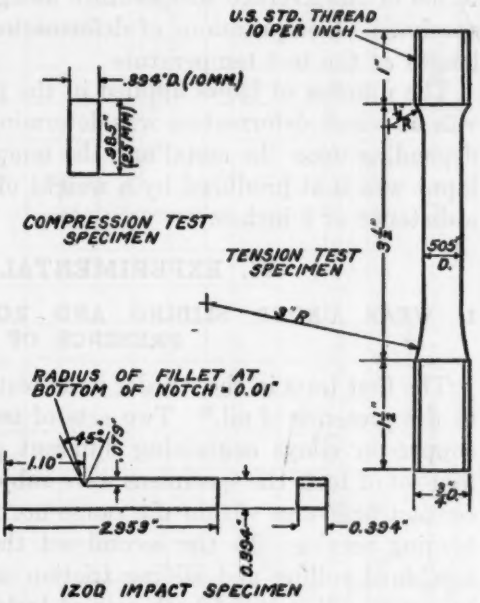

FIG. 16.—*Test specimens used in the mechanical tests*

In making tests at elevated temperatures a preheating furnace and an equalizing furnace (*E*, fig. 18) were used to expedite the work, as in the case of the notched-bar impact tests. Specimens were kept

[22] French, H. J., "Methods of test in relation to flow in steels at various temperatures," Proc. Am. Soc. Test. Matls., **26**, pt. 2. p. 7; 1926. Also see Bureau of Standards Letter Circular 238; Nov. 11, 1927.

[23] The authors are indebted to E. M. Staples, research associate for the Bunting Brass & Bronze Co., Toledo, Ohio, for the design and construction of this pounding machine.

at the test temperature in the preheating furnace, removed as required, and placed in the equalizing furnace in which they remained, in this case, throughout the test.

The deformation of the specimen under the selected number of blows was measured with an ordinary micrometer in tests at atmospheric temperatures and by an optical micrometer, by sighting through a window in the heating furnace, in tests at elevated temperatures. Temperatures were measured with a potentiometer and 22-gauge chromel-alumel thermocouples and were controlled from a thermocouple mounted in the anvil after this had been calibrated in terms of the average temperature along the central axis of the test specimen. Computations of deformation were based on the original length at the test temperature.

The number of blows applied in the pounding tests and the intervals at which deformation was determined varied in individual tests, depending upon the metal and the temperature of test. The energy input was that produced by a weight of 7.15 pounds falling through a distance of 2 inches.

VI. EXPERIMENTAL RESULTS

1. WEAR UNDER SLIDING AND ROLLING FRICTION IN THE PRESENCE OF OIL

The first tests in the Amsler wear-testing machine were carried out in the presence of oil.[24] Two sets of tests were made on a group of copper-tin alloys containing different proportions of lead. In the first set of tests the specimens were subjected to sliding friction under contact pressures within the range normally encountered in railroad bearing service. In the second set the bronzes were subjected to combined rolling and sliding friction under relatively high contact pressures. Results of both sets of tests are summarized graphically in Figures 20, 21, and 22.

A steel disk was rotated against the bronze specimen for a selected number of revolutions and the wear determined by loss of weight. Repeated and consecutive determinations were made with the same pair of specimens (bronze and steel) until a uniform wear rate was attained.

As shown in Figures 20, 21, and 22, the amount and rate of initial wear varied widely for the different bronzes, but after an initial period the wear-revolution and wear-work curves turned toward the horizontal axis. After a preliminary "wearing-in" period, the wear rate dropped to very low values in the presence of a lubricant, and it appeared impracticable to develop with any degree of certainty

[24] The oil used (Mobiloil A) had the following properties: Viscosity at 210° F. = 58; viscosity at 100° F. = 577; flash point = 375° F.; union color number = 8; pour point below 20; neutralization number = 0.44; carbon residue = 0.26 per cent. These properties were determined by H. T. Kennedy, associate scientist.

differences between the different alloys. Therefore, it seemed necessary to make wear tests without lubrication to throw additional light upon the wearing properties of these bronzes.

The described results appear to be consistent with the view that after an initial period in which the contact surfaces "wear-in"

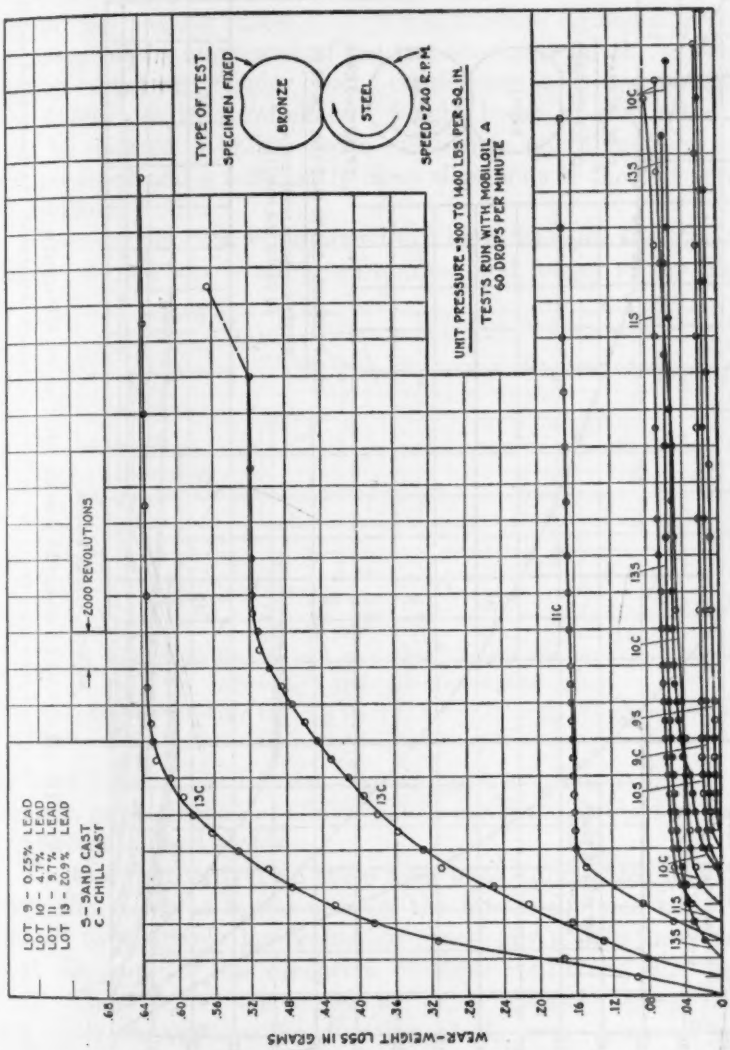

FIG. 20.—*Wear-revolution curves for bronzes tested under lubrication*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. They were tested against steel A.

bearings will undergo little or no further wear so long as complete lubrication is maintained. Therefore, considerable attention was given to the behavior of the different metals in wear tests without lubrication.

However, some useful information concerning the different alloys was secured from the tests made in the presence of oil. Examination

of Figures 20, 21, and 22 shows that increase in lead in the copper-tin-lead alloys was accompanied by a general increase in the amount of wear and of work in the initial or "wearing-in" period and likewise by an increase in the slip (or revolutions) comprising this period.

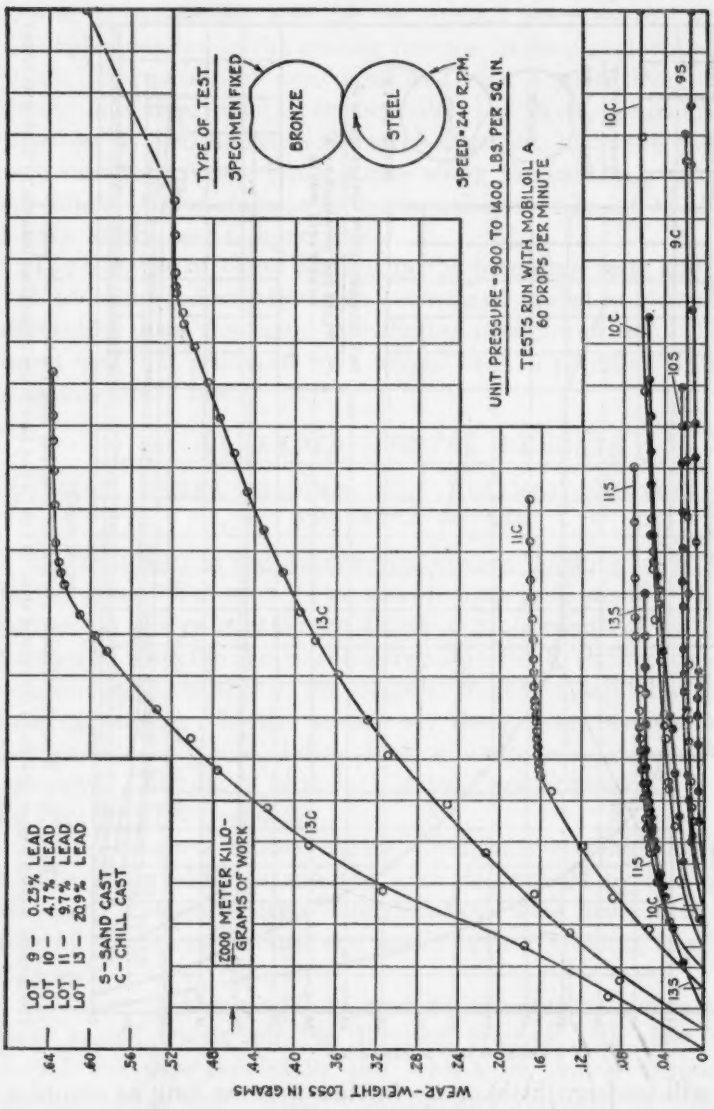

FIG. 21.—*Wear-work curves for bronzes tested under lubrication*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. They were tested against steel A.

These effects are summarized graphically in Figure 23, but the curves should not be interpreted too rigorously due to the marked influence upon the total wear and the wear rate in the early stages of operation of the initial condition of the surfaces of the specimens. These were finish ground by ordinary shop methods and there were

quite probably variations in the roughness of the test surfaces at the start of the tests.

It should be recognized that the magnitude and duration of the "wearing-in" period will also be dependent upon the lubricant and the relation of this lubricant to the initial condition of the surface of the bearing metal. They will likewise be dependent on the accuracy of the alignment of the test specimens in the Amsler wear testing machine. Under such conditions it is conceivable that the differences in duration and weight losses of the initial period may be reduced or obliterated if sufficiently smooth contact surfaces are provided and a sufficiently close alignment of the test specimens is obtained.

However, the data summarized in Figure 23 show that the "wearing-in" period was much more marked and longer in the high-lead

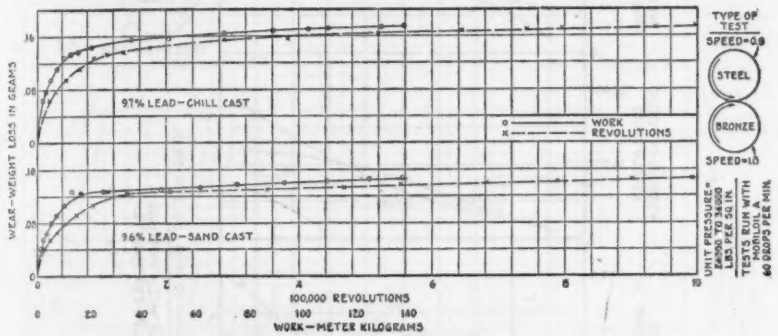

FIG. 22.—*Wear-revolution and wear-work curves for bronzes tested under lubrication and high-contact pressure*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5. They were tested against steel A.

bronzes than in the bronzes containing 0.25 per cent lead. It was likewise more marked in alloys cast in metal molds than in the same alloys cast in sand molds.

The torque records from these wear tests summarized in Figure 24 show that after an initial interval the torque dropped and "settled down" to relatively low values in practically all of the copper-tin-lead alloys with the exception of those containing 0.25 per cent lead. These latter continued to show widely varying frictional properties throughout the duration of the wear tests. Thus, the copper-tin alloys with 0.25 per cent lead may be considered somewhat inferior to the high-lead bronzes from the standpoint of frictional properties but, as already indicated, have an advantage over the latter in having a shorter "wearing-in" period with less loss in weight.

In Figure 25 is shown a summary of the wear per unit of work in relation to the total slip. Since there are no standard methods of

interpreting the results of wear tests, Figure 25 is included to show that an added method of examining the experimental data confirms some of the important features emphasized in the discussion of the preceding charts.

The weight changes in the steel specimens for the tests under sliding friction in the presence of oil are shown in Figure 26. In no case did the bronzes cause appreciable wear on the soft steel specimens.

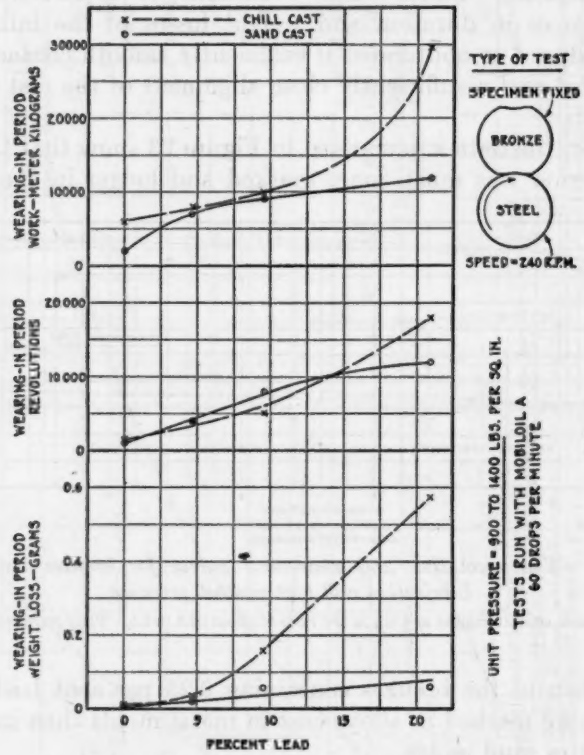

FIG. 23.—*Effect of lead on the "wearing-in" period of bronzes tested under lubrication in the Amsler wear-testing machine*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5. They were tested against steel A.

Some of the bronze test specimens were examined under the microscope subsequent to the wear tests to determine the degree of deformation of the metal at the contact surfaces. In no case was evidence of work hardening found in the specimens subjected to rolling and sliding friction in the presence of oil. Two of the specimens which were subjected to sliding friction in the presence of oil showed strain lines as illustrated in Figure 27, *b*. Surface flow was observed in a high-lead bronze (fig. 27, *a*) but not in any of the low-lead bronze specimens examined.

2. WEAR TESTS AT ORDINARY TEMPERATURES IN THE ABSENCE OF OIL

Wear tests under sliding friction without oil were attended by difficulties even at fairly low contact pressures. The temperature of the bronze specimens was quickly raised and resulted in a marked flow of the surface metal. Under such conditions there was difficulty in determining just what constituted wear and how much wear took place, and there was likewise a question regarding the exact conditions

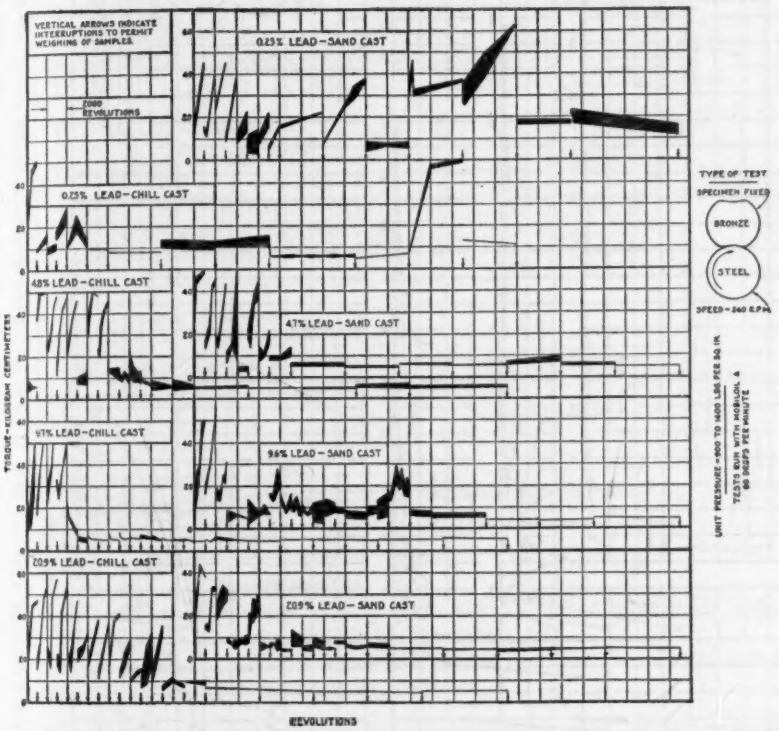

FIG. 24.—*Torque in testing different bronzes against axle steel A in the presence of oil in the Amsler wear-testing machine*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. The width of the black lines represents torque fluctuations.

under which the tests were made. Variations in the frictional and thermal properties of the different metals probably produced variations in the temperatures reached and direct comparisons of the experiments did not seem justified.

Such difficulties were minimized with the lower rate of slip and the small area of contact between the two disks in tests under combined rolling and sliding friction. There was less heat developed in a given time, despite the relatively high unit contact pressures, and there was probably also a better opportunity for dissipation of the heat resulting

from the frictional forces. Therefore, all wear tests without lubricants which are described in this report were made under combined rolling and sliding friction.

(*a*) FILM FORMATION.—The first tests were characterized by erratic weight losses; that is, individual specimens did not show con-

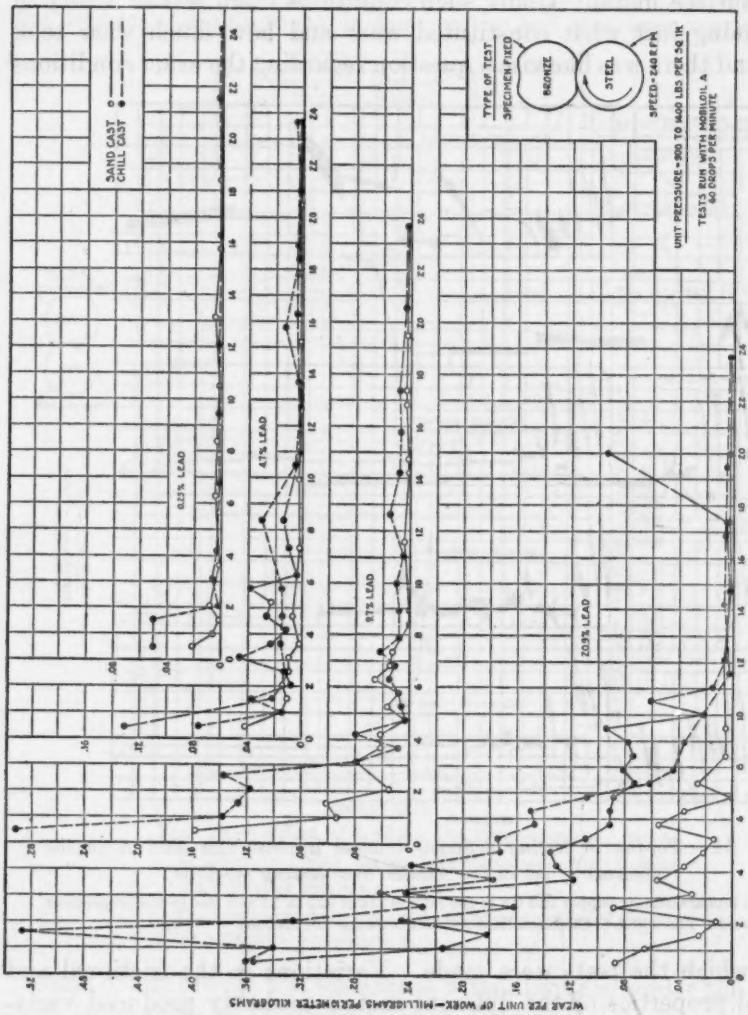

FIG. 25.—*Relation between wear per unit of work and slip for bronzes tested under lubrication*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and the proportions of lead indicated. They were tested against steel A.

sistent rates of wear in consecutive tests, even after a preliminary period. In addition to a change in the roughness of the contact surfaces from the original ground finish, a change in color and what appeared to be a coating or film was frequently observed. Typical films are illustrated in Figures 28, *i*, 30, *c*, and 30, *d*.

The formation of such coatings which were not present at the start of the tests appeared to be related to changes in the wearing and

frictional properties of the metals. The compositions of the films produced are not now known with certainty, but they were apparently of two types. The first, comprising films which looked like copper oxide, was observed in the copper-tin alloys which were low in lead. These same films also occurred in the high-lead bronzes, but they were not so easy to detect in the latter alloys, due to the more frequent formation of a second type of coating, a black film which was either lead or a lead compound. Tests so far made, including X-ray analysis, have failed to reveal its exact nature.

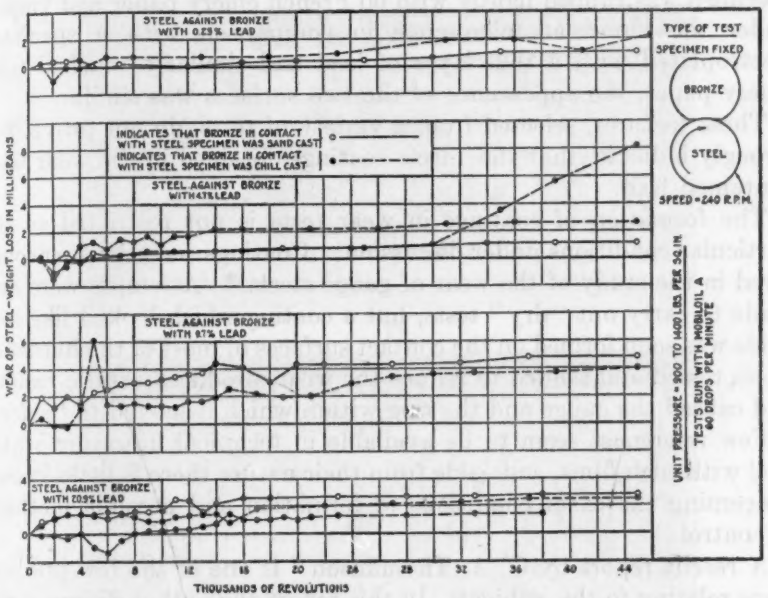
STEEL AGAINST BRONZE WITH 0.23% LEAD
INDICATES THAT BRONZE IN CONTACT WITH STEEL SPECIMEN WAS SAND CAST
INDICATES THAT BRONZE IN CONTACT WITH STEEL SPECIMEN WAS CHILL CAST
STEEL AGAINST BRONZE WITH 4.1% LEAD
STEEL AGAINST BRONZE WITH 8.7% LEAD
STEEL AGAINST BRONZE WITH 20.9% LEAD
WEAR OF STEEL—WEIGHT LOSS IN MILLIGRAMS
THOUSANDS OF REVOLUTIONS
TYPE OF TEST
SPECIMEN FIXED
BRONZE
STEEL
SPEED = 240 R.P.M.
UNIT PRESSURE = 300 TO 1400 LBS. PER SQ. IN.
TESTS RUN WITH MOBILOIL A
60 DROPS PER MINUTE

FIG. 26.—*Wear of axle steel A in contact with different bronzes in the wear tests made in the presence of oil*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and the proportions of lead indicated. Points below 0 represent "loading" of the steel.

Diffraction patterns obtained [25] in the examination of a bronze specimen containing about 15 per cent of lead are reproduced in Figure 31. Pattern *b* is for a specimen with the black coating, while pattern *a* is for a similar specimen without film. One of the significant differences between the two is that the lines from the specimen with the film have only about one-fifth the intensity of those from the specimen without the film. This difference is probably due to the partial absorption of the X rays by the black coating which reduced the energy available for diffraction by the base metal copper-tin solution. A similar result was obtained with change in casette position and location. The fact that the surface film itself yielded no

[25] By F. Sillers, jr., junior metallurgist.

pattern may be due to the high degree of fragmentation, distortion, and preferred orientation to which the small quantity of material constituting the film was subjected during the wear test

Chemical analysis of a small particle taken from the surface of a black-coated specimen gave a positive test for lead, but it was not certain that the surface layer alone was being tested. When heated in the oxidizing portion of a Bunsen flame, a bronze specimen "sweated" black globules which looked like lead and which gave a lead diffraction pattern under X-ray examination. When a coated specimen was rubbed lightly with 00 French emery paper and viewed under the binocular microscope in comparison with a specimen electroplated with a thin layer of lead and similarly touched with emery paper, the appearance of the two surfaces was similar.

These features, selected from a variety of tests, do not prove, but strongly indicate, that the black coatings obtained in the wear tests contained lead.

The formation of coatings in wear tests is not restricted to the particular conditions under discussion. Coatings have been encountered in the study of the wear of gauge steels.[26] Attempts were first made to carry out "dry" tests, but a coating which looked like iron oxide was soon formed on the contact surfaces of most of the hardened steels tested and tended to reduce the wear, produced erratic results, and caused the gauge and the ring within which it moved to "seize."

Few references seem to be available in technical literature which deal with such films, and aside from their nature there is little known concerning the exact conditions of formation and possible methods of control.

A recent report by G. A. Thomlinson[27] is one of the few publications relating to this subject. In this report the author discusses the conditions of formation of iron oxide when a steel ball is moved in contact with other metallic surfaces.

Aside from the exact nature of such films and the conditions of formation, it appeared certain from the preliminary unlubricated wear tests of bronzes that film formation played a part in the erratic results of the wear tests, and that some method of elimination or control of films should be provided if consistently uniform results were to be obtained.

Wear is ordinarily considered to take place through the displacement and detachment of small particles from the contact surfaces. The energy accompanying the tearing off of such particles can be conceived to appreciably raise the temperature of small masses to

[26] French, H. J., Herschman, H. K., Wear of steels with particular reference to plug gauges, Tr. Am. Soc. Steel Treat., **10**, No. 5, p. 683; November, 1926.

[27] Thomlinson, G. A., "The rusting of steel surfaces in contact." Proc. Roy. Soc., Series A, **115**, No. A771, p. 472; July 1, 1927.

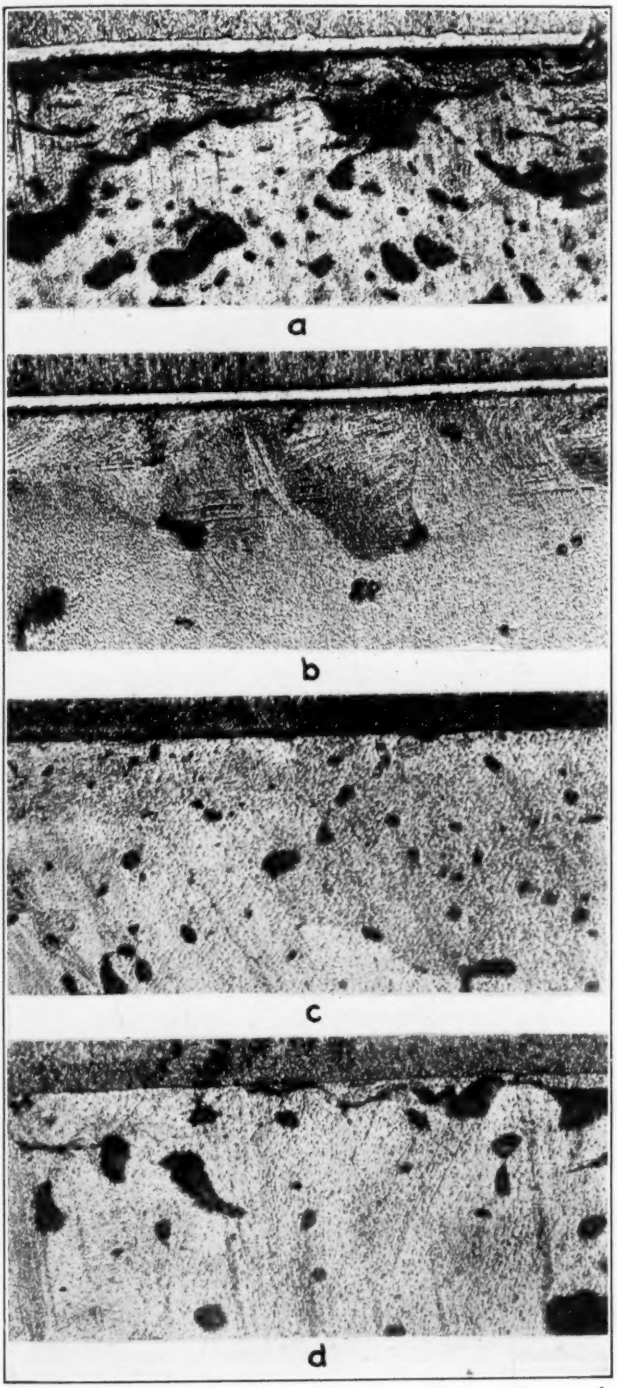

FIG. 27.—*Micrographs of the structure adjacent to the wearing surfaces of four bronzes containing copper and tin in the ratio of about 92.5 to 7.5. The top layers in each picture are copper or copper and nickel plate.* ×250

Etched with solution of $FeCl_3$ plus HCl: *a*, Bronze containing 20.9 per cent lead showing surface flow under lubricated sliding friction; *b*, Bronze containing 0.25 per cent lead showing strain lines at wearing surface. Lubricated sliding friction; *c*, Bronze containing 4.7 per cent lead showing no sign of surface workhardening. Lubricated sliding friction; *d*, Same bronze as *a*, but tested under combined rolling and sliding friction

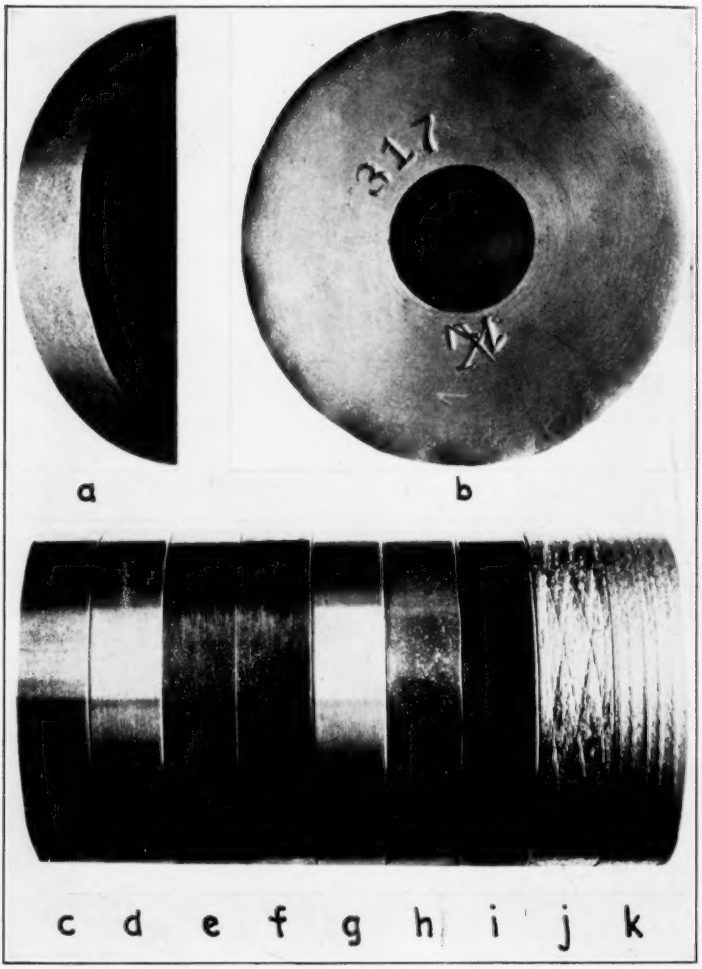

FIG. 28.—*Some of the wear test specimens before and after test (approximately full size)*

a, "Flat-wheel" condition in one of the bronze specimens
b, Another view of specimen *a*
c, Steel specimen before test (machined surface)
d, A bronze specimen before test (machined surface)
e, Bronze specimen showing rough type of wear
f, Bronze specimen showing the "flat-wheel" condition
g, A bronze specimen before test (machined surface)
h, Bronze specimen showing smooth type of wear
i, Bronze specimen showing film
j, Bronze specimen showing ridges
k, Steel specimen showing ridges

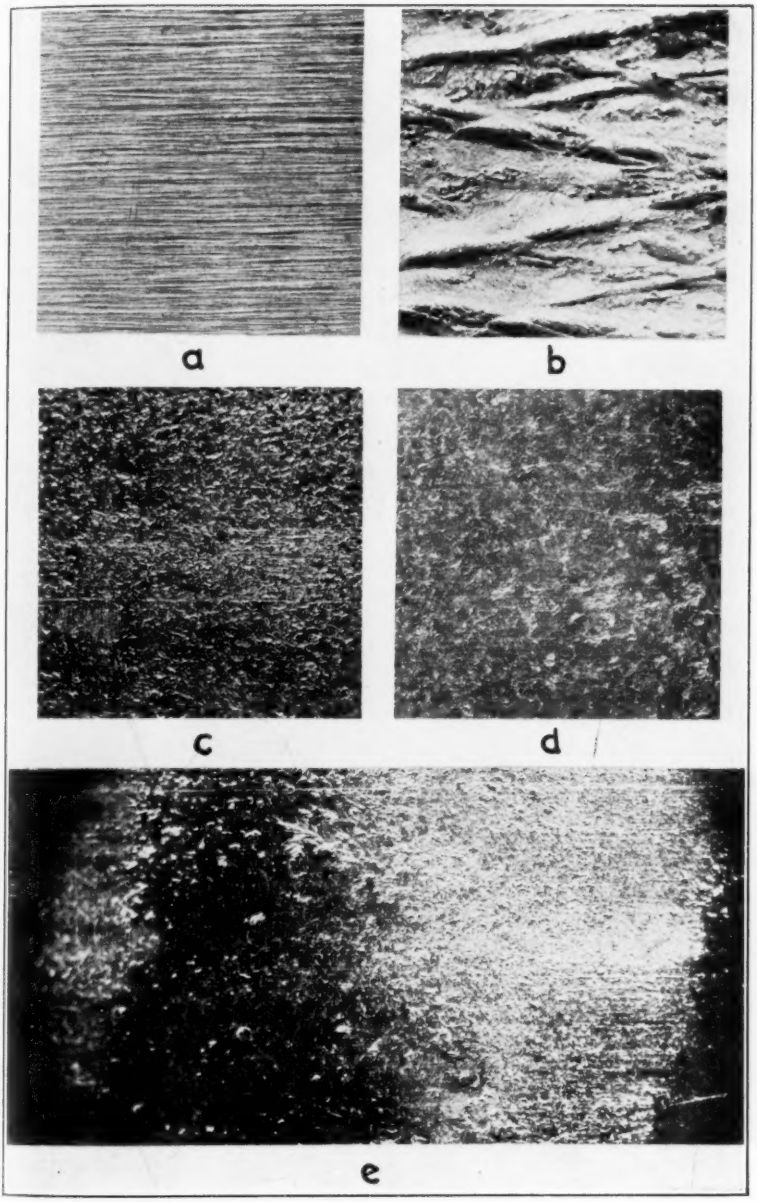

FIG. 29.—*Surfaces of some of the bronze specimens before and after wear tests.* ×10

a, Original machine-finished surface
b, Very rough with ridges
c, Rough type of wear
d, Smooth type of wear
e, "Flat-wheel" condition

UNIV. OF MICH.

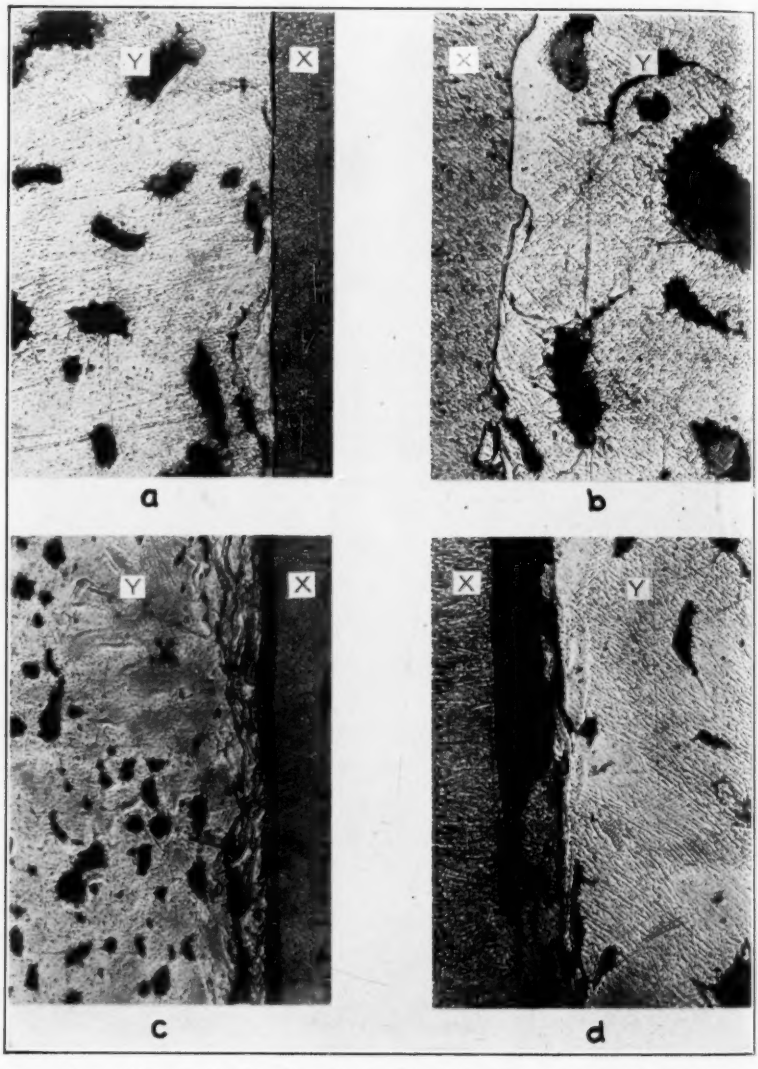

FIG. 30.—*Micrographs showing the metal adjacent to worn surfaces of some of the bronze wear test specimens.* ×250

Etched with solution of 3 parts NH_4OH plus 1 part H_2O_2
x, Represents copper plate used to preserve the contour of the bronze surface
y, Represents the bronze specimen
a, Shows smooth type of wear
b, Shows rough type of wear
c, Shows film on surface with underlying fragmented bronze
d, Shows heavy film combined with rough type of wear

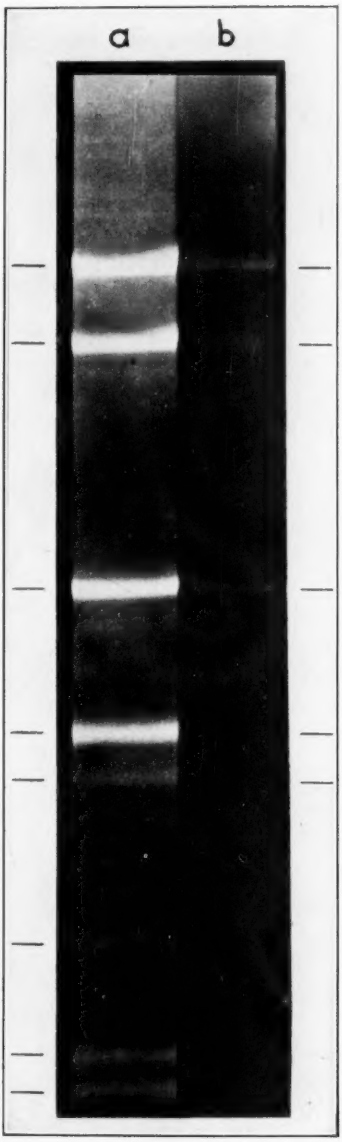

FIG. 31.—*X-ray diffraction patterns of bronze specimens with and without black film*

a, Without film *b*, With film

UNIV. OF MICH.

the point of rapid oxidation. In at least some cases the films formed appeared to consist of embedded oxidized particles. This was most noticeable in the tests of gauge steels previously mentioned, but

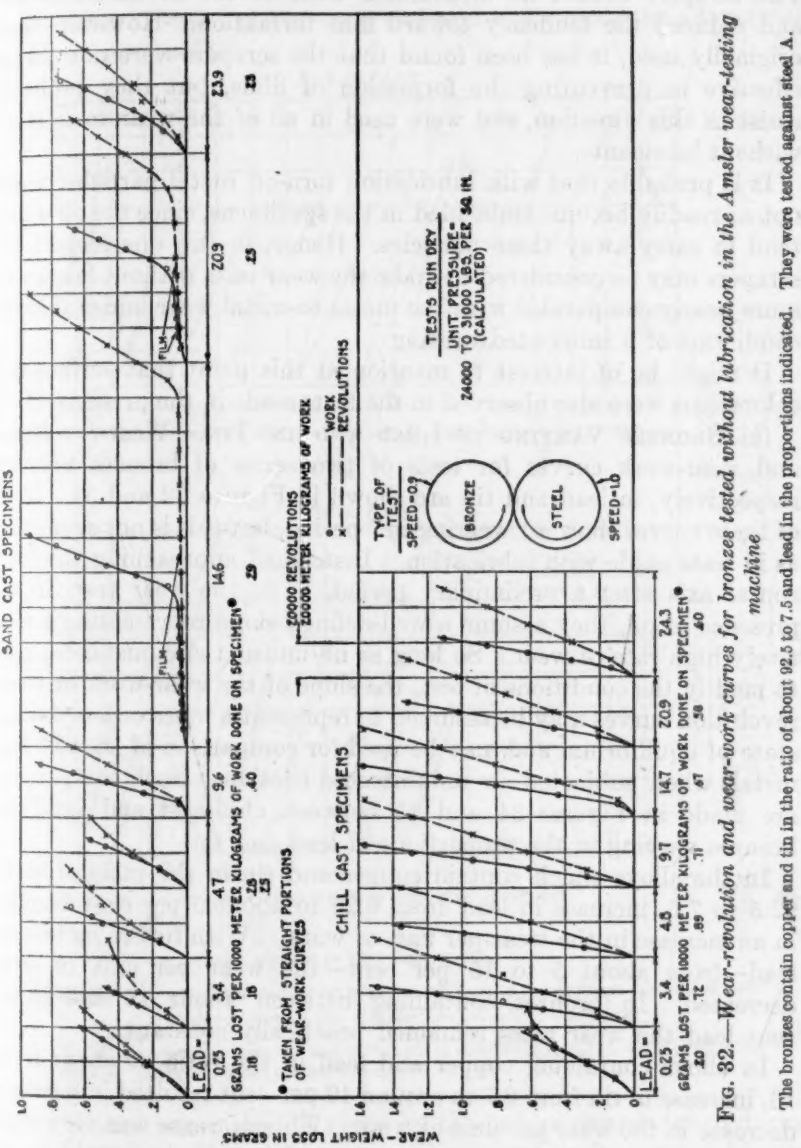

FIG. 32.—*Wear-revolution and wear-work curves for bronzes tested without lubrication in the Amsler wear-testing machine*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. They were tested against steel A.

was sometimes clearly shown on bronze specimens having films which looked like copper oxide. Removal of abraded particles as soon as these became detached might reasonably be expected to result in a smaller tendency toward film formation.

Based on this line of reasoning, scrapers were installed on the Amsler wear testing machine. These consisted of half-hard copper strips which continually rubbed against each of the rotating disks. The scrapers caused no measurable wear of the bronzes or steels and reduced the tendency toward film formation. However, since originally used, it has been found that the scrapers were not always effective in preventing the formation of films, but they probably assist in this direction, and were used in all of the wear tests made without lubricants.

It is probable that with lubrication torn-off metal particles would not so readily become embedded in the specimens, since the oil would tend to carry away these particles. Hence, in this one respect the scrapers may be considered to make the wear tests without lubricants more nearly comparable with the metal-to-metal wear under starting conditions of a lubricated bearing.

It might be of interest to mention at this point that surface discolorations were also observed in the tests made in the presence of oil.

(*b*) Bronzes Varying in Lead and in Tin.—Wear-revolution and wear-work curves for tests of two series of bronzes varying, respectively, in lead and tin are shown in Figures 32 and 33. Most of these curves show a "wearing-in" period, but this is not so marked as in tests made with lubrication. Instead of approaching the horizontal axis after a preliminary period, as in the wear tests in the presence of oil, they assume a well-defined slope representing a relatively high rate of wear. So long as no unusual circumstances arise to modify the conditions of test, the slope of the wear-work or wear-revolution curves may be assumed to represent a more or less definite state of equilibrium and may be used for comparison of the different metals when subjected to unlubricated friction. Such comparisons are made in Figures 34 and 35 between chill-cast and sand-cast bronzes varying in the proportions of lead and tin.

In the alloys which contain copper and tin in the ratio of about 92.5 to 7.5, increase in lead from 0.25 to about 5 per cent resulted in an increase in the wear per unit of work. With futher increase in lead—from about 5 to 15 per cent—the wear per unit of work decreased. In bronzes containing between about 15 and 25 per cent lead the wear rates remained practically constant.

In alloys containing copper and lead in the ratio of about 84 to 16, increase in tin from 0.7 to around 10 per cent resulted in a general decrease in the wear per unit of work. This decrease was very rapid between 0.7 and about 3 per cent tin and became less rapid as the proportion of tin was further raised to 10 per cent.

Similar changes were observed when considering the wear per unit of slip (or wear per 10,000 revolutions) and, therefore, no separate summaries of these data have been included in this report.

In general, alloys cast in metal molds wore much more rapidly than corresponding alloys cast in sand molds. Two exceptions were found in which the wear per unit of work was about the same for the chill-cast and sand-cast conditions. These were the copper-lead mixture (with 0.7 per cent tin) and the copper-tin alloy (with 0.25 per cent lead).

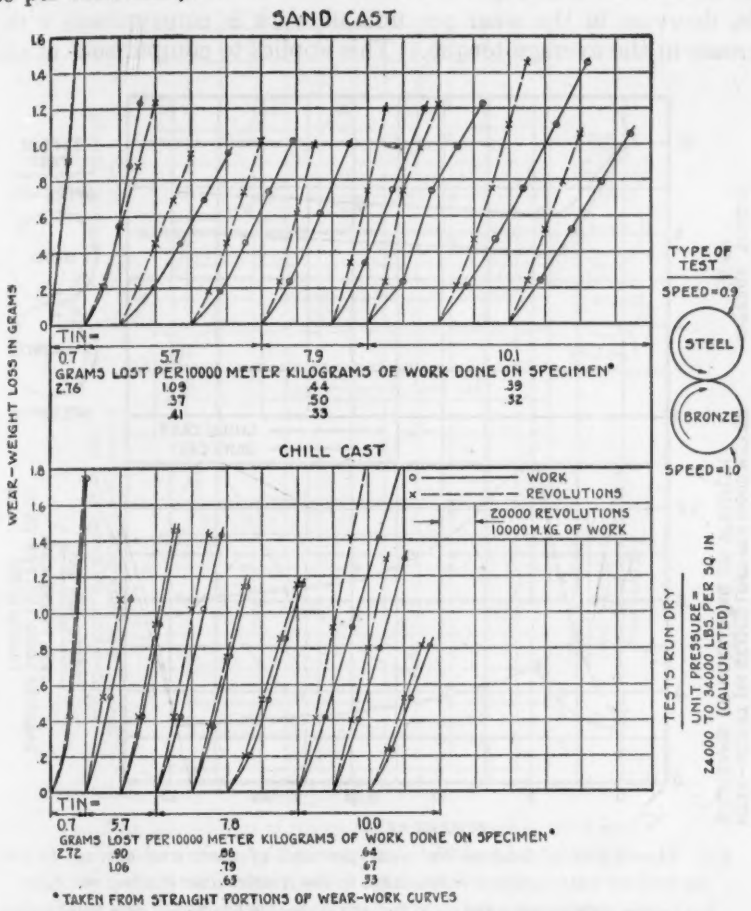

FIG. 33.—*Wear-revolution and wear-work curves for bronzes tested without lubrication in the Amsler wear-testing machine*

The bronzes contain copper and lead in the ratio of about 84 to 16 and tin in the proportions indicated. They were tested against steel A.

Additional information is given by the torque records taken during the wear tests. Typical torque-revolution curves for a set of bronzes varying in tin are shown in Figure 36. Low torque values were obtained at the start in the majority of the wear tests, but as the bronze and steel specimens continued to rotate in contact the torque increased. In many cases it approached a more or less definite and

high value, while in others it quickly reached a maximum and then gradually decreased throughout the wear-test interval. In general, however, the torque records from consecutive wear tests of any bronze specimen were quite similar and permit comparisons of average values for each set of consecutive wear tests. Such comparisons are included in Figures 34 and 35 and show that, as a general rule, decrease in the wear per unit of work is concomitant with an increase in the average torque. This applies to comparisons of alloys

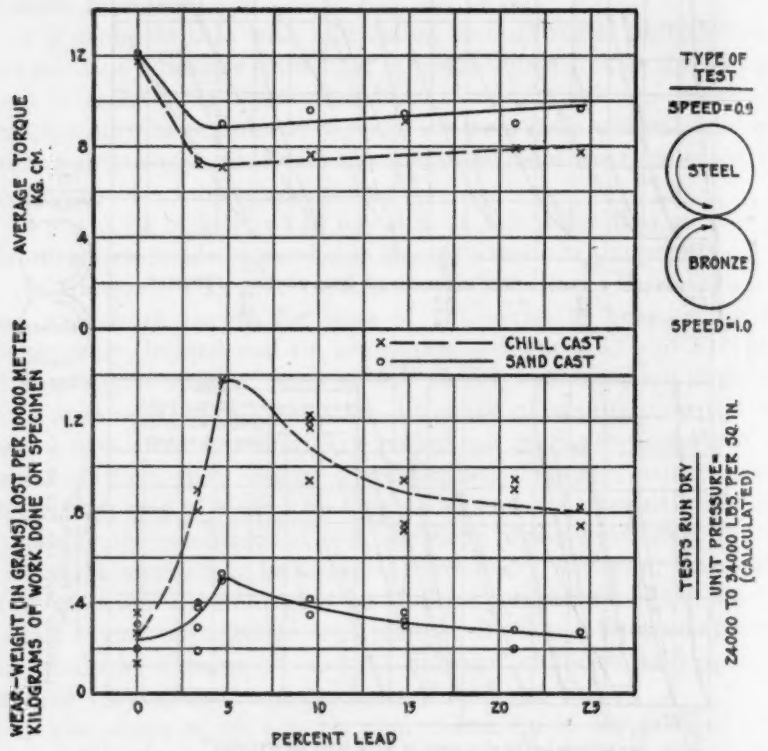

FIG. 34.—*Effect of lead on the wear per unit of work and average torque for bronzes tested without lubrication in the Amsler wear-testing machine*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and were tested against steel A.

varying in tin or lead and likewise to comparisons between the chill-cast and sand-cast conditions of a given composition. Exceptions are the low-tin alloys which assumed a very rough surface condition in the wear tests and showed erratic torque values in duplicate specimens.

Since the "wearing-in" period in these wear tests without oil was relatively small and the initial rise in torque was restricted to a small portion of the test interval, the foregoing comparisons develop the interesting feature that where the average torque is high the wear

per unit of work (also wear per revolution), subsequent to the "wearing-in" period, is frequently low. There is no obvious explanation for these effects which will, however, be given further consideration in a subsequent section of this report.

Examination of the surfaces of the test specimens throughout the progress of the tests likewise revealed differences between the dif-

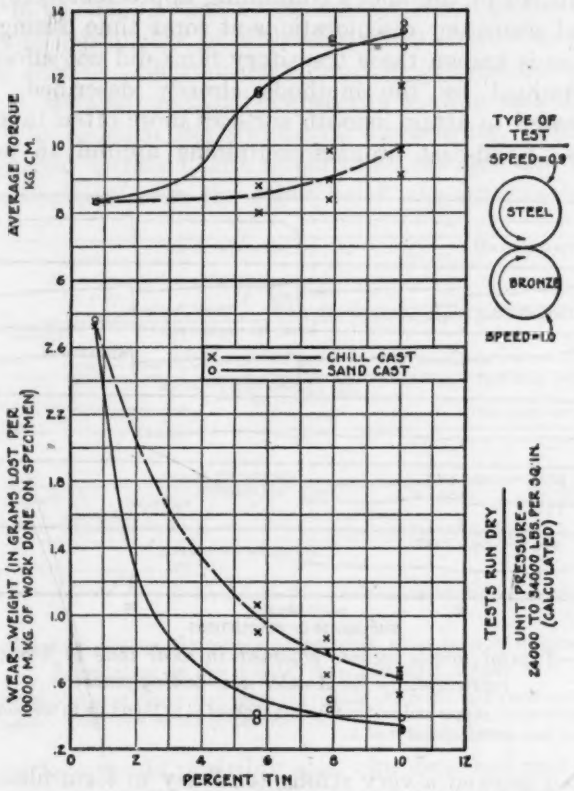

FIG. 35.—*Effect of tin on the wear per unit of work and average torque for bronzes tested without lubrication in the Amsler wear-testing machine*

The bronzes contain copper and lead in the ratio of about 84 to 16 and were tested against steel A. The inflection around 2.5 per cent tin in the torque curve for sand-cast bronzes is based on additional data included in Figure 45. The form of the wear curves below 5.5 per cent tin is approximate for reasons given in Section VI, 2 (*c*), of this report.

ferent alloys. Some of the surfaces became exceedingly rough and showed the ridges illustrated in Figures 28, *j*, and 29, *b*. Others became moderately rough, as shown in Figures 28, *h*, 29, *c*, and 30, *b*, while in some cases films were formed as illustrated in Figures 28, 30, *c*, and 30, *d*. A "flat-wheel" effect was also observed and is illustrated in Figures 28, *a*, 28, *b*, 28, *f*, and 29, *e*.

Surface roughening accompanied by ridges was quite largely restricted to the copper-lead mixture low in tin and the copper-tin alloys low in lead. These also more quickly reached the "flat-wheel" condition than the remaining alloys in the groups tested. When the "flat-wheel" condition became marked, heavy pounding resulted in the wear tests.

Practically all of the alloys containing appreciable proportions of lead showed transitory discolorations at some time during the tests, but as far as is known these transitory films did not affect the wear rates determined by the methods already described. Sand-cast alloys appeared to attain smooth surfaces more often than the chill-cast alloys. Sand-cast bronzes containing around 15 per cent or

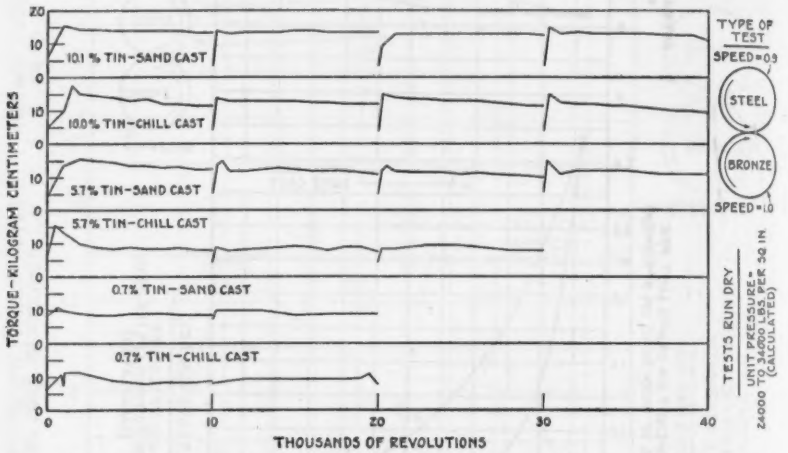

FIG. 36.—*Typical torque records obtained in wear tests of bronzes without lubrication in the Amsler wear-testing machine*

The bronzes contain copper and lead in the ratio of about 84 to 16 and tin in the proportions indicated. They were tested against steel A.

more of lead showed a very strong tendency to form films which did not break down in prolonged tests. These changes in surface condition can readily be seen by comparing the different photographs with the original machine-finished surface illustrated in Figure 29, *a*.

In one or two cases fragmentation of the metal was observed immediately below the surface of the worn test specimen. An interesting example of this is shown in Figure 30, *c*, where the fragmented bronze is just below a dark surface film. This film is thin in comparison with that shown over the roughened surface illustrated in Figure 30, *d*.

Where surface films were produced, the wear rate usually decreased materially, and there was generally a marked decrease in the average torque. The reduction in wear rate accompanying film formation is illustrated in Figure 32, while the decreased torque accompanying both transitory films and films maintained for appreciable intervals is

illustrated in Figure 37. Could the described films be produced when desired and also maintained, they might be of decided advantage in practical service because they would reduce the wear of the bearings and also decrease the frictional power losses when lubrication failed.

The copper-tin alloys containing 0.25 per cent lead showed the lowest wear rates, but they quickly assumed a rough surface condition (illustrated in fig. 29, *b*) when subjected to friction in the absence of oil. This would probably be disadvantageous from the standpoint of practical service. Short interruptions in lubrication which permitted direct metallic contact between the bearing and shaft might be expected to roughen the surface of the bearing. The recesses in the roughened surface might retain the lubricant, and so tend to maintain lubrication. On the other hand, the irregularities of the

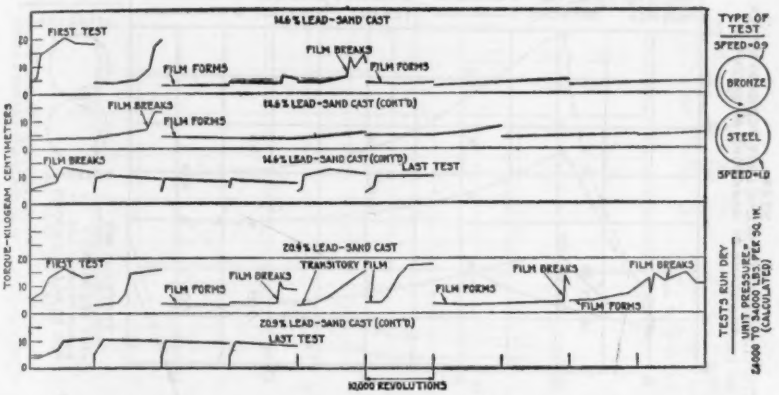

FIG. 37.—*Effect of film on the torque observed in wear tests of bronzes without lubrication in the Amsler wear testing machine*

The two bronzes contained copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. They were tested against steel A.

roughened surface would probably also trap detached metallic particles and any nonmetallic abrasives which might be present and so foster the scoring of the shaft. Thus, the entrapped lubricant would be mixed with abrasive particles, and so might promote wear of both the bearing and shaft.

Another important characteristic of the bronzes low in lead is that they showed much higher torque than the bronzes containing from 10 to 25 per cent of lead. During interruptions in lubrication increased torque accompanying the roughening of the surface of the bearing would tend to raise the frictional temperatures. Maintenance of metallic contact for any appreciable time might, therefore, result in a marked increase in operating temperature which, in turn, would alter the properties of the bronze and might readily be conceived to promote, if not cause, seizure. High torque, likewise, means high power consumption in the bearings, which is also disadvantageous.

(*c*) DATA SHOWING THAT THE WEAR TEST RESULTS CAN BE DUPLICATED.—In view of the emphasis already placed upon the complexity of the problem of wear testing, the question arises whether the described differences in the wearing properties of the different bronzes would be duplicated if the original experiments were repeated. The original wear tests on the group of bronzes containing different proportions of lead, which are summarized in Figure 34, were made during March and neighboring months in the year 1926 by an operator who will be designated as A. About a year later, during February, 1927, specimens from the same castings (not necessarily the same disks) were turned down to somewhat smaller diameters, and a new operator, designated as B, was assigned the work of making a com-

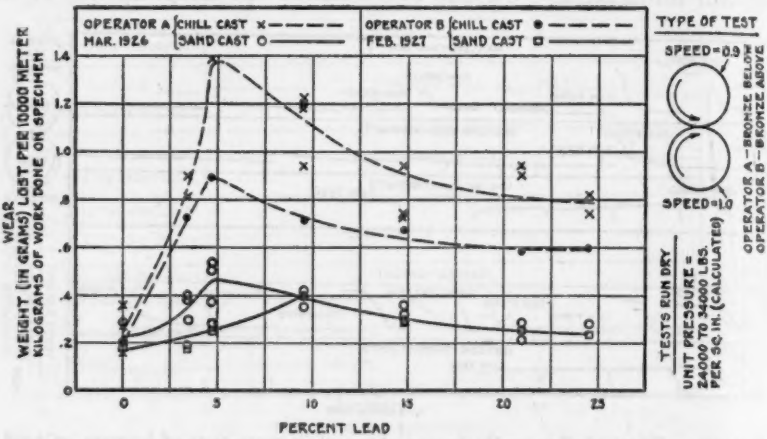

FIG. 38.—*Comparison of two independent sets of wear tests without lubrication made about a year apart by two different operators on bronzes varying in lead*

See text for detailed discussion. The bronzes contain copper and tin in the ratio of about 92.5 to 7.5.

plete set of tests comparable to the original. Operator B had had a limited experience with the Amsler wear-testing machine but had taken no part in the first set of experiments. He was instructed to interchange the positions of the bronze and steel specimens so that instead of having the bronzes on the lower spindle of the testing machine, as in the first set of tests, they would be mounted on the upper spindle.

The results obtained are included in Figure 38 with those obtained in the first set of tests made about a year earlier. Despite the change in position of the specimens, the use of different disks from various castings, and a change in the operator, close checks were obtained. The lower wear per unit of work in the chill-cast bronzes in the second set of tests is due to the fact that smaller disks were used. The wearing surfaces of the disks were, therefore, further from the metal which had been chilled to the greatest degree, and hence wore less rapidly.

The differences in the wear rates in the first and second sets of tests for the sand-cast alloys containing about 2.5 to 10 per cent of lead is likewise due to a variation in the degree of chill of the metal. In one case a sand-cast specimen, which gave a relatively high rate of wear under conditions comparable to those of the first set of tests, was turned down and then gave a lower wear rate which checked the value obtained on the smaller disks used in the second set of wear tests.

Additional evidence is given in Figure 39 that the results obtained in the unlubricated wear tests can be reproduced. Two sets of experiments were made on bronzes varying in tin, but in this case the work was done by one operator. The original tests were repeated after a period of about six months, and, as shown in Figure 39, the

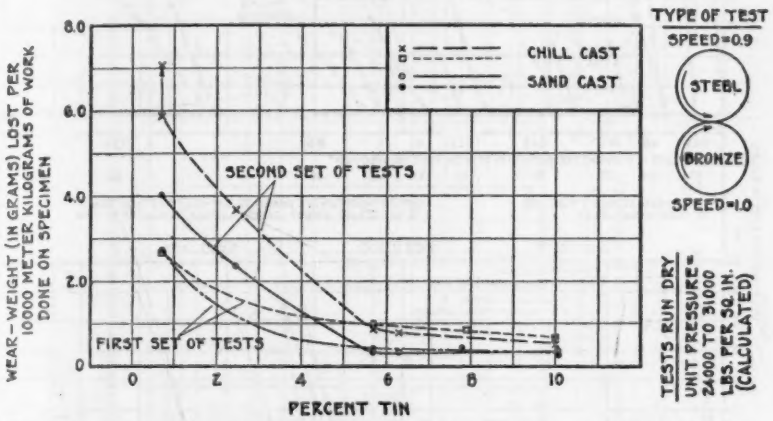

FIG. 39.—*Comparison of two independent sets of wear tests without lubrication made at an interval of six months on bronzes varying in tin*

See text for detailed discussion. The bronzes contain copper and lead in the ratio of about 84 to 16.

numerical values of the wear per unit of work are in fairly close agreement, except for the alloys containing around 2.5 per cent or less of tin. However, the general form of the curves shown in Figure 39 is the same in both sets of experiments.

The observed differences in the wear of the low-tin bronzes in the two sets of experiments may be explained by the fact that these alloys showed very high wear and quickly attained the "flat wheel" condition previously described. Only a few consecutive tests could be made before appreciable pounding was obtained in the Amsler wear-testing machine. It was, therefore, impracticable to make a sufficient number of consecutive tests for the wear-work curves to assume definite slopes. In other words, the preliminary period was not passed within the limits of the experiments, and the wear per unit of work calculated from the results of the last test interval was greater than the average

wear rate for the last few tests. Under such conditions relatively large differences may be expected in duplicate tests. Duplicate samples do not, ordinarily, "wear-in" at exactly the same rates and at the end of a fixed number of revolutions will be at different stages of the "wearing-in" or preliminary period.

The wear rate in the last test interval for the low-tin alloy (chill-cast) is indicated by the arrow in Figure 39 to permit comparison

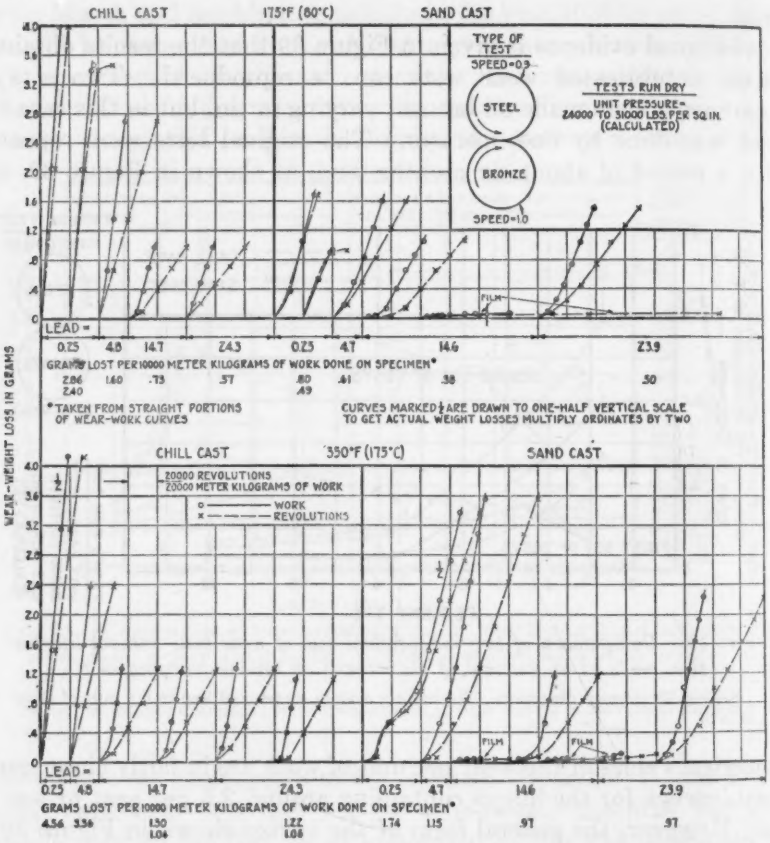
CHILL CAST
175°F (80°C)
SAND CAST
TYPE OF TEST
SPEED=0.3
STEEL
BRONZE
SPEED=1.0
TESTS RUN DRY
UNIT PRESSURE=
24000 TO 31000 LBS. PER SQ. IN. (CALCULATED)
FILM
LEAD =
0.25 4.8 14.7 24.3 0.25 4.7 14.6 23.9
GRAMS LOST PER 10000 METER KILOGRAMS OF WORK DONE ON SPECIMEN*
2.86 2.40 1.60 .73 .57 .80 .49 .41 .38 .30
*TAKEN FROM STRAIGHT PORTIONS OF WEAR-WORK CURVES
CURVES MARKED ½ ARE DRAWN TO ONE-HALF VERTICAL SCALE TO GET ACTUAL WEIGHT LOSSES MULTIPLY ORDINATES BY TWO
WEAR-WEIGHT LOSS IN GRAMS
CHILL CAST
350°F (175°C)
SAND CAST
20000 REVOLUTIONS
20000 METER KILOGRAMS OF WORK
o ——— WORK
x - - - - - REVOLUTIONS
½
FILM
LEAD =
0.25 4.8 14.7 24.3 0.25 4.7 14.6 23.9
GRAMS LOST PER 10000 METER KILOGRAMS OF WORK DONE ON SPECIMEN*
4.56 3.36 1.30 1.04 1.22 1.08 1.74 1.15 .91 .91

FIG. 40.—*Wear-revolution and wear-work curves for bronzes tested at different temperatures without lubrication*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5 and lead in the proportions indicated. They were tested against steel A.

with the average wear rates for the last few tests which are represented by the curves. The higher wear rate in the last interval does not, however, change the general form of the curves in Figure 39.

3. WEAR TESTS AT ELEVATED TEMPERATURES IN THE ABSENCE OF OIL

(*a*) BRONZES VARYING IN LEAD.—Typical wear-revolution and wear-work curves from tests at elevated temperatures are shown in

Figure 40, and resemble those obtained from tests at atmospheric temperatures. Sand-cast bronzes containing about 15 to 25 per cent of lead again showed a strong tendency toward film formation which was, perhaps, more marked than at atmospheric temperatures. In several cases represented in Figure 40 these films did not break down under prolonged test, and after a short initial period the specimens showed practically no wear. Therefore, duplicate samples

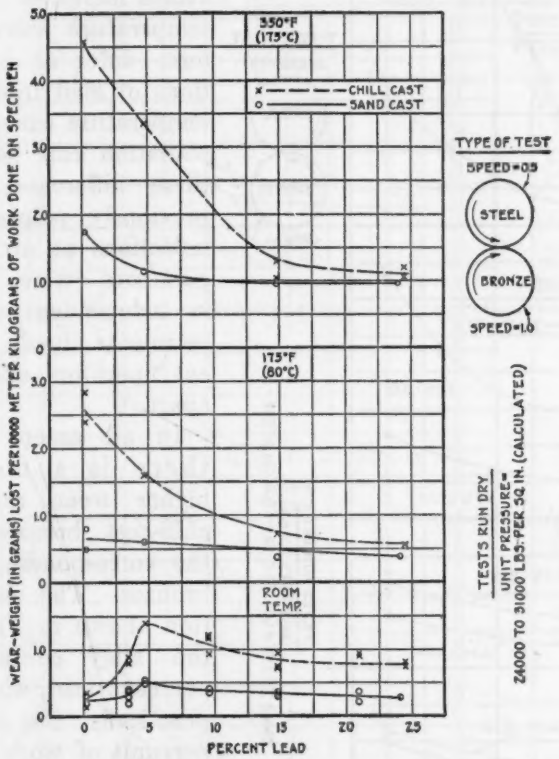

FIG. 41.—*Effect of lead on the wear per unit of work for bronzes subjected to friction without lubrication at different temperatures*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5. They were tested against steel A.

were tested and, as a general rule, no such permanent films were encountered. In addition to low-wear rates accompanying film formation, there was generally a decrease in the average torque.

The effects of variations in lead on the wearing properties of both chill-cast and sand-cast bronzes at different temperatures are shown in Figure 41. At 350 and 175° F. (175 and 80° C.) the wear per unit of work decreases rapidly from 0.25 to about 12 or 15 per cent lead but shows only minor changes with further increase to 25 per cent lead.

Above about 5 per cent lead the room temperature curves have the same general form as those for tests at elevated temperatures; with decrease in lead below about 5 per cent the wear rate at atmospheric temperatures decreases, while that at the elevated temperatures increases. The room temperature curves, therefore, differ at low proportions of lead from the high temperature curves. No explanation can be given for these differences, but, as previously pointed out, the inflections at ordinary temperatures were duplicated in independent sets of experiments made by different operators about a year apart.

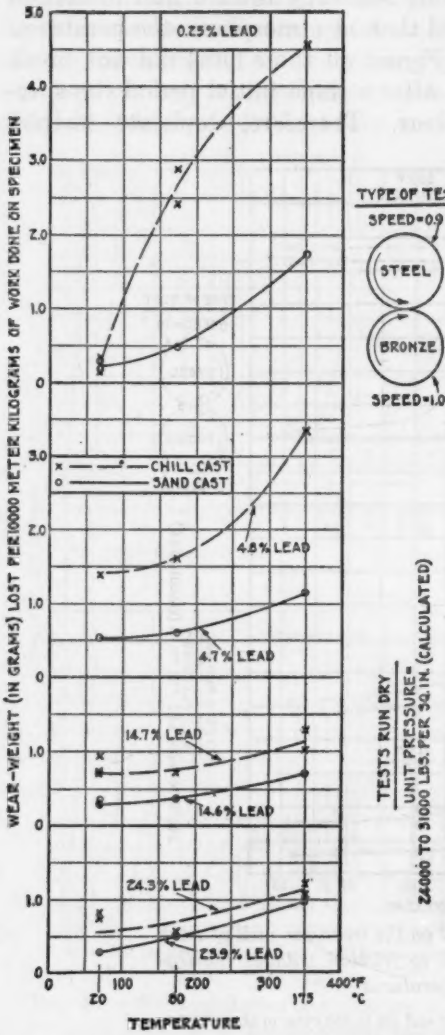

FIG. 42.—*Effect of temperature on the wear per unit of work of bronzes varying in lead when subjected to friction without lubrication in the Amsler wear testing machine*

These are the same tests referred to in Figure 41.

In all except one case there is a measurably higher wear rate in the chill-cast bronzes than in the corresponding sand-cast bronzes. The single exception shown in Figure 41 is the alloy containing 0.25 per cent lead, which showed practically the same wear per unit of work at atmospheric temperatures in the chill-cast and the sand-cast conditions. The chill-cast bronzes also appear to "wear in" more rapidly than the corresponding sand-cast bronzes.

Figure 41 shows that the large differences in wearing properties of the bronzes are between the low-lead and the high-lead alloys, and that variations in lead between about 15 and 25 per cent do not materially affect the wear rates.

Figure 42 shows that rise in temperature increases the wear rate, but that this increase is greater in bronzes containing about 5 per cent or less of lead than in those with 15 to 25 per cent of lead.

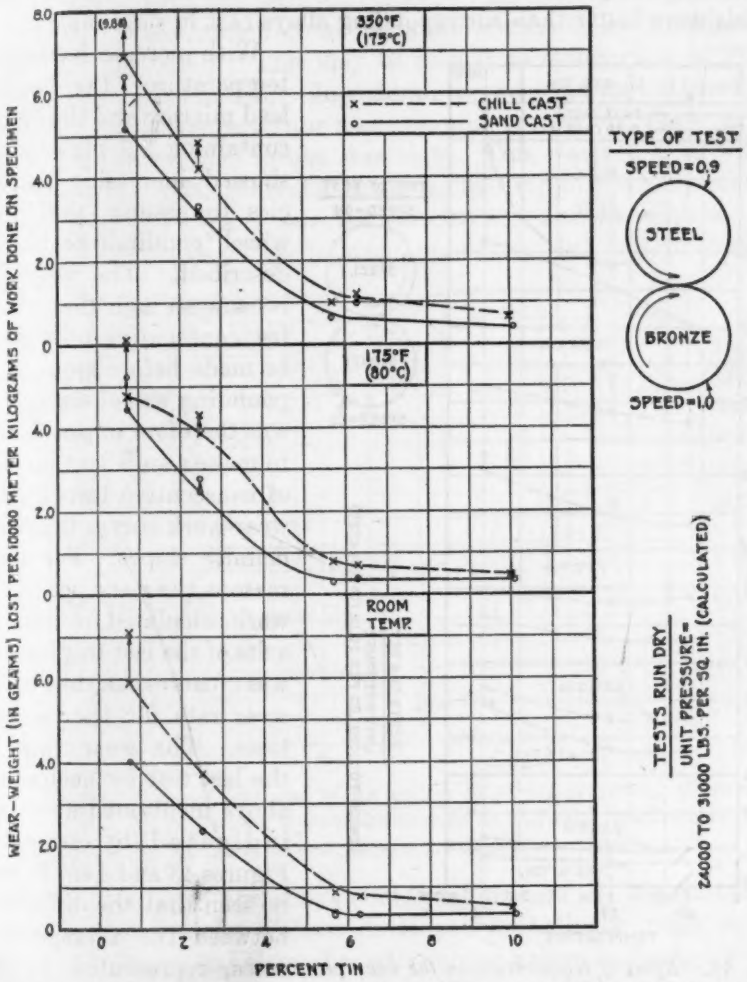

FIG. 43.—*Effect of tin on the wear per unit of work for bronzes subjected to friction without lubrication at different temperatures*

The bronzes contain copper and lead in the ratio of about 84 to 16. They were tested against steel A at atmospheric temperatures and against steel B at other temperatures.

(*b*) BRONZES VARYING IN TIN.—The effects of tin on the wearing properties of copper-tin-lead alloys at different temperatures are summarized in Figure 43. The wear per unit of work decreases appreciably with increase from 0.7 to about 5 per cent tin in both the chill-cast and sand-cast specimens; further increase from about 5 to 10 per cent tin does not appreciably affect the wear rates at any

of the test temperatures. However, as shown in Figure 44, increase in the wear rate produced by increase in temperature above atmospheric is less in the bronze containing 10 per cent tin than in bronzes containing 2.5 per cent or less of tin. In all cases alloys cast in metal molds wore faster than corresponding alloys cast in sand molds.

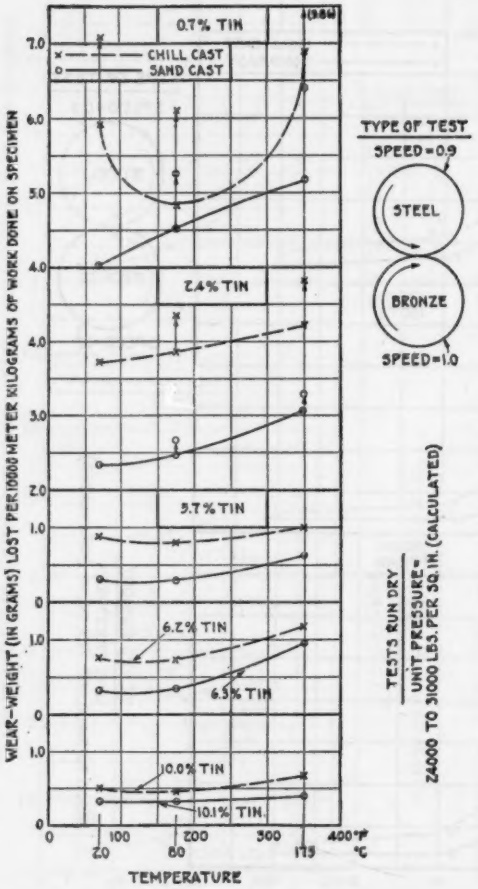

FIG. 44.—*Effect of temperature on the wear per unit of work of bronzes varying in tin when subjected to friction without lubrication in the Amsler wear testing machine*

These are the same tests referred to in Figure 43.

With increase in the test temperatures the copper-lead mixture and the bronze containing 2.5 per cent tin showed increasing tendencies to assume the "flat-wheel" condition previously described. The wear also became so high that only a few consecutive tests could be made before appreciable pounding was observed. It was therefore impracticable to make a sufficient number of consecutive tests for the wear-work curves to assume definite slopes. For these reasons the wear per unit of work calculated from the results of the last test interval was greater than the average wear rate for the last few tests. The wear rates in the last test for each of the alloys mentioned above are indicated by arrows in Figures 43 and 44. It may be seen that the differences between the average wear rates, represented by the curves, and the wear rate in the final test, represented by the arrows, become greater as the test temperature is increased and as the tin content of the bronzes is decreased. However, in no case does the somewhat higher wear rate in the last test change the conclusions which were drawn from the average rates of wear represented by the curves in Figures 43 and 44.

The foregoing comparisons, which are based on the wear per unit of work, are confirmed by comparisons based on the wear per unit of

slip. Therefore, separate summaries of the wear per unit of slip are not included in this report.

(*c*) FRICTIONAL PROPERTIES *v.* WEAR.—Comparisons of the average torque and wear in the high temperature wear tests developed effects similar to those observed in tests at atmospheric temperatures previously described. This may be seen from comparison of Figure 45 with Figures 41 and 43. In general, the wear per unit of work was high when the average torque was low, and, conversely, the wear was low when the average torque was high. This was not always true, but was found quite generally in the comparison of the chill-cast and sand-cast alloys of any selected composition and likewise in comparison of the bronzes containing appreciably different proportions

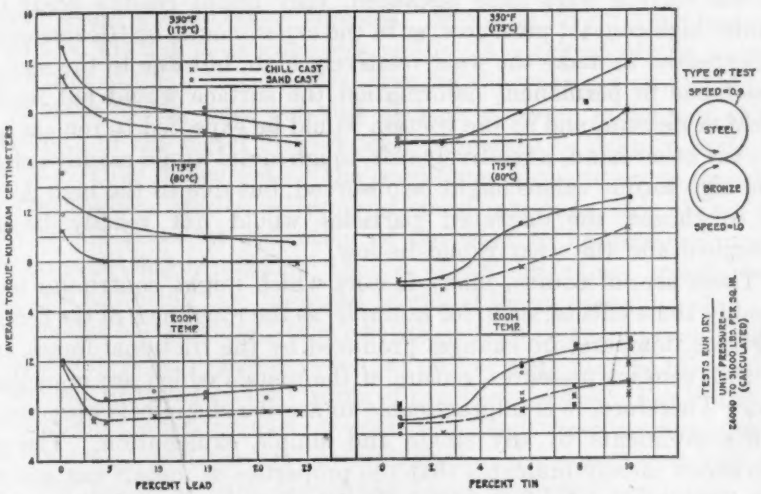

FIG. 45.—*Effect of lead and of tin on the frictional properties of bronzes subjected to friction without lubrication at different temperatures*

These data were obtained in the tests referred to in Figures 41 to 44, inclusive.

of lead or of tin. The alloys containing around 2.5 per cent or less of tin are exceptions. These developed exceedingly rough surfaces in the wear tests and showed erratic torque values in duplicate specimens, so that the curves are only indicative of the general trends at the low tin side of the diagrams.

As indicated in a previous section of this report, there is no obvious explanation for such effects which might, at first glance, be considered unusual. However, the cases where high wear is concomitant with relatively low average torque are so numerous in the described experiments that it seems important to consider possible reasons for such behavior.

The torque records obtained from the pendulum dynamometer on the Amsler wear testing machine represent the total resistance to

slip; that is, the total of the frictional forces at the contact surfaces of the two metals. Wear is conceived to take place by the displacement and detachment of small particles from the contact surfaces, and, as has already been pointed out, the resistance to displacement and detachment depends not only on the resistance of the metals to permanent deformation (hardness) but also on toughness.

If two alloys are considered which differ in toughness and in their resistance to permanent deformation (hardness) so that the one may be characterized as hard and brittle while the other is relatively soft and tough, differences in the wearing and frictional properties may be expected. The alloy which is hard and brittle would be expected to offer high resistance to permanent deformation, but when particles at the surface were once displaced, they might readily break off. Under high contact pressures, as in the experiments, brittleness would be expected to make the wear relatively high, but due to the marked resistance to permanent deformation the surface would not readily yield or deform, and so the friction would be expected to remain low. On the other hand, a relatively soft, tough metal would readily deform and high torque values might be observed, but due to the high degree of toughness, the displaced particles would not readily become detached and the wear would be low.

There are, of course, many factors which might contribute to or modify these effects; such, for example, as the roughness of the contact surfaces, temperature changes produced by the frictional forces, rate of slip, contact pressures, nature of the metals which are in contact, etc. Therefore, it is impracticable to ascribe the effects observed in the experiments to any single and simple explanation. The one advanced merely indicates that the properties of metals can account for high torque being associated with low wear, but high torque combined with high wear might be observed with changes in one or more of the conditions of test. Furthermore, the assumed differences in hardness and toughness upon which the foregoing explanation is based were certainly much greater than those existing in the alloys tested.

The question of how best to express the wear rate was given considerable thought. Generally, other investigators have reported wear in terms of distance traveled. This does not give a complete picture. Several factors which are known to have a very marked effect upon the wear are entirely independent of the distance traveled by a wear-test specimen. Such variables as the condition of the test surfaces, the presence or absence of film, the normal pressure, all of which are known to influence the wear, are entirely independent of revolutions of the test roller. True, in evaluating wear tests upon the basis of revolutions some of these factors can be fixed and so noted, but others, such as the condition of the contact surfaces, quite frequently change

continuously throughout a test. This change is reflected in the amount of wear and friction but not in the distance traveled by the test specimen.

In evaluating wear as a function of work, an effort is made to correlate the wear with a physical expression which not only reflects the physical conditions of test but which is also a measure of the actual energy required to produce a certain amount of wear under certain conditions. Such factors as pressure, condition of the test surfaces, etc., are reflected in the frictional force existing between the test surfaces, and changes in these factors produce corresponding changes in the friction. To calculate the energy consumed in friction during a test, the distance traveled by the test specimen must also be used, since work is force (the friction existing between the specimens) times the distance traveled. Thus, in evaluating wear as a unit of work a more complete picture is obtained of the wearing properties of the metals than is given by the wear per unit of slip.

In other words, wear per unit of work gives an indication of the resistance of the metal to wear when subjected to definite frictional forces. While influenced by the conditions of operation, it does not express results directly in terms of speed, pressure, etc. Wear per unit of slip gives an indication of the wear to be expected under certain operating conditions; that is, certain pressures, speeds, etc., and does not bring into view the frictional forces encountered.

It should be recognized that conditions may be encountered where comparisons based solely on wear per unit of work may be misleading or difficult to translate into terms of practical service. For example, low values of wear per unit of work may result from moderate wear per revolution and moderately high frictional properties (torque in the wear tests) or from somewhat lower wear per revolution and relatively low torque. Metals which showed only moderately low values of wear per unit of work through the latter combination of wear and frictional properties would have an advantage over metals which showed the lowest wear per unit of work produced through the former combination since low wear and low friction are both important from the standpoint of bearing service. Hence, until a more complete understanding is obtained of the relation between frictional properties and wearing properties of metals it is advisable, to consider wear, work, and distance traveled as well as other features, such as the character of the worn surfaces which have been discussed in this report.

Attention has already been called in Section VI, 2, (*b*) to the fact that the changes in wear per unit of work produced by variations in chemical composition and methods of casting were similar to those in the wear per unit of slip (revolutions) for the groups of alloys con-

sidered. It was therefore unnecessary to include graphical summaries both of the wear per unit of work and the wear per unit of slip.

(*d*) WEAR OF THE STEELS.—As a general rule, the weight losses were erratic for the steel specimens which were rotated in contact with the bronzes. Therefore, it was impracticable to make comparisons of the wear per unit of work or the wear per unit of slip. However, the different bronzes produced different effects upon the steels against which they were tested, and these effects may be illustrated in a general way by comparing the total wear of the steels in the first 40,000 revolutions of the bronze specimens. Where films were formed which temporarily reduced the wear, the values selected for comparisons represent the weight losses of the steels for the

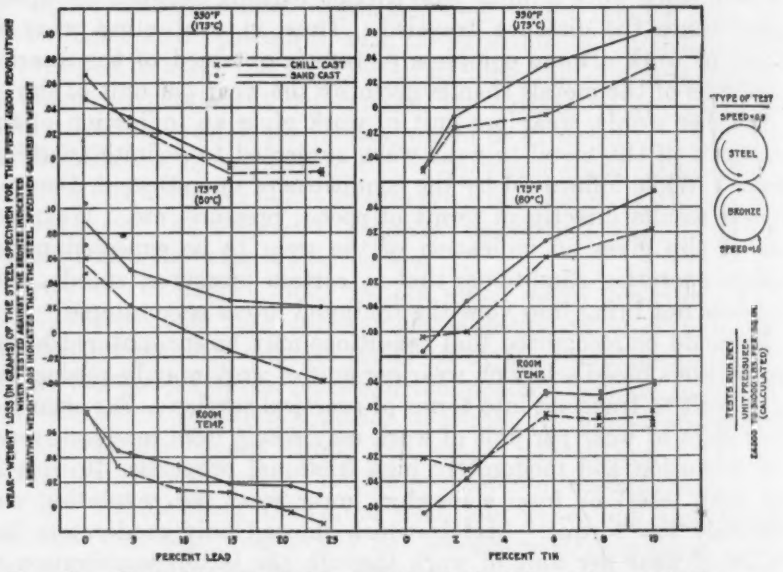

FIG. 46.—*Wear of the axle steel in contact with the different bronzes in the wear tests without lubrication*

first 40,000 revolutions of the bronze subsequent to the breakdown of the film. These comparisons should not be interpreted too rigorously, because it was rarely possible to remove entirely the particles from the bronzes which were deposited on the steel specimens. The data in Figure 46 are intended only to give a general idea of differences observed in the tests.

Examination of Figure 46 shows three particularly interesting features, as follows: (1) As the lead content of the bronzes was raised the wear of the steel decreased. With the high-lead bronzes the steel specimens frequently showed no wear but, on the contrary, increased in weight. This increase was probably due to the fact that the deposit of particles from the bronze on the steel was large in compari-

son with the wear of the steel. (2) As the tin content of the bronzes was raised, the wear of the steel specimens increased. (3) With but few exceptions the wear of the steels was greater when in contact with sand-cast bronzes than when used against the corresponding chill-cast alloys.

The bronzes which caused the highest wear of the steels would naturally be expected to score shafts in practical service much more quickly than those which had the smallest effect upon the steel specimens in the described tests. Therefore, in this respect, there is a distinct advantage to the high-lead bronzes.

4. SAND ABRASION TESTS

The form of the curves obtained by plotting the total weight of sand blasted upon the specimens against the total loss in weight were quite similar to the wear-work curves obtained under unlubricated friction in the Amsler machine; that is, after a short initial period, the weight loss was directly proportional to the amount of sand used. The results of the sand-blast tests are summarized in Figure 47.

In all cases but one the sand-cast alloys were more resistant to the sand abrasion than the corresponding chill-cast alloys. The one exception was the bronze containing 0.25 per cent lead which showed practically the same wear in the chill-cast and sand-cast conditions. This is similar to the effect observed in the Amsler unlubricated wear tests on the same alloys.

As shown in Figure 47, increase in lead increased the rate of wear in the bronzes containing an approximately constant ratio of copper to tin. Furthermore, the numerical values of the wear of the high-lead bronzes are exceedingly high in comparison with the wear of the bronze containing 0.25 per cent lead. The lead content is therefore an important factor in the resistance of these alloys to sand abrasion. The lead particles which are mechanically held in the matrix are probably torn out, and because of the high specific gravity of lead the weight losses are large.

Figure 47 also shows that increase in tin from 0.7 to 10 per cent decreases the wear in the alloys containing an approximately constant ratio of copper to lead. In these alloys the proportion of lead decreased as the proportion of tin increased. As already indicated, decrease in lead resulted in a greatly improved resistance to the sand abrasion; therefore, the decrease in wear with increase in tin which is shown by the curves in the upper part of Figure 47 is due to the benefits of lowered lead as well as to the beneficial effects of increased tin.

With one major exception the sand-blast abrasion tests placed the alloys in the same general order as that given by the Amsler

wear tests without oil. In both types of test alloys cast in metal molds wore more rapidly than the corresponding alloys cast in sand molds. Likewise, wear resistance was improved by increase in tin from 0.7 to about 5 per cent; between 5 and 10 per cent tin the changes in wearing properties were small.

Variations in lead produced opposite effects in the Amsler wear tests and the sand-abrasion tests. In the former increase in lead

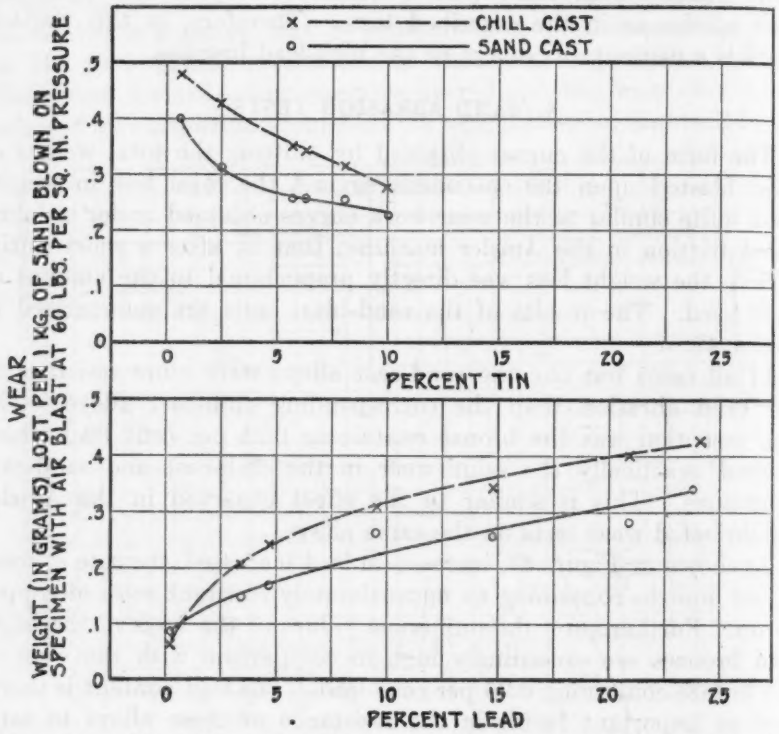

FIG. 47.—*Effect of lead and of tin on the wear of bronzes subjected to sand abrasion*

The alloys varying in lead contained copper and tin in the ratio of about 92.5 to 7.5; those varying in tin contained copper and lead in the ratio of about 84 to 16.

improved resistance to wear while in the latter it decreased wear resistance.

5. SINGLE-BLOW IMPACT TESTS

The results of the impact tests are summarized graphically in Figures 48 to 51, inclusive. With the exception of the alloys low in lead, the resistance to impact in the Izod test was very low. As shown in Figure 48, increase in lead from 0.25 to 25 per cent produced a progressive decrease in impact resistance at all temperatures between atmospheric and 600° F. (315° C.). The magnitude of this decrease was generally greater with lead increase from 0.25 to about 10 per

cent than from 10 to 25 per cent; it was likewise greater at 350° F. (175° C.) and lower temperatures than at 600° F. (315° C.).

Figure 49 shows that the impact resistance increased with increase in tin from a trace to about 5 per cent and then decreased with increase from around 5 to 10 per cent tin. These changes were of somewhat greater magnitude in the sand-cast bronzes than in the

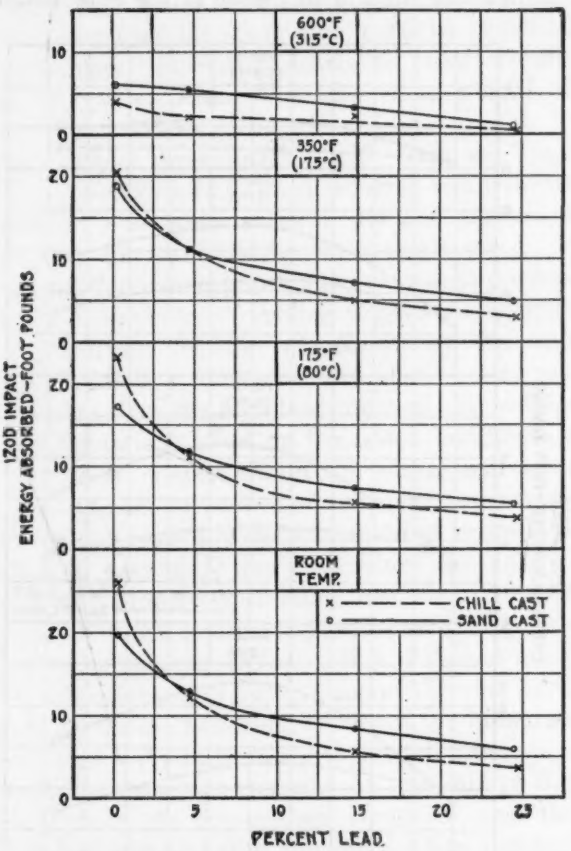

FIG. 48.—*Effect of lead on the notched-bar impact resistance of bronzes at different temperatures*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5.

chill-cast bronzes; they were likewise more marked at 350° F. (175° C.) and lower temperatures than at 600° F. (315° C.).

Figures 10 and 11 show that the eutectoid first appeared in the castings with around 7.5 per cent tin. So long as the limit of solid solution of tin in copper is not exceeded, an increase in the tin content improves the impact resistance of the bronzes (fig. 49), but additions producing appreciable amounts of the eutectoid appear to lower the impact values.

With one exception, the sand-cast bronzes showed about equal or greater resistance to impact than the corresponding chill-cast bronzes. This was the bronze with 0.25 per cent lead in which the chill castings showed a better impact resistance than the sand castings at all temperatures except 600° F. (315° C.).

Chill-cast bronzes have a finer grain size, smaller lead particles, and, when sufficiently high in tin, more of the eutectoid than corre-

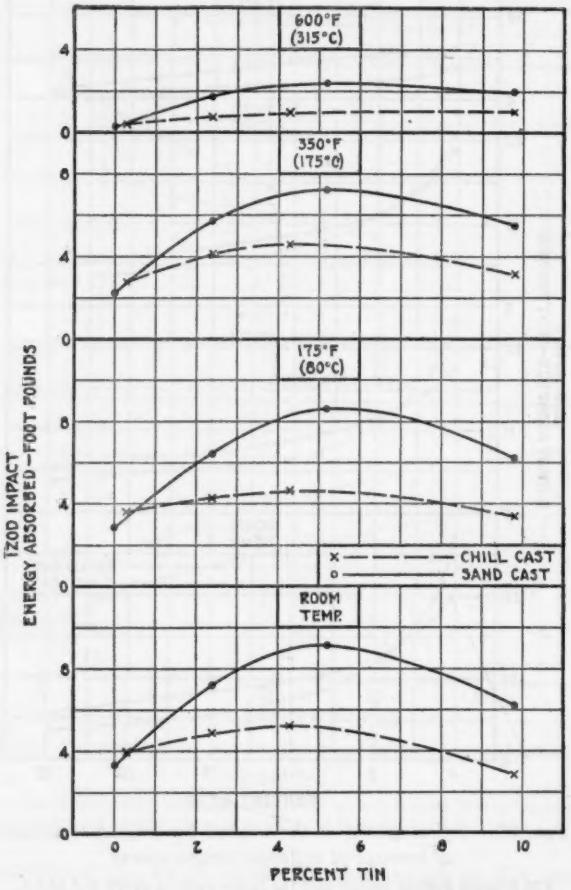

FIG. 49.—*Effect of tin on the notched-bar impact resistance of bronzes at different temperatures*

The bronzes contain copper and lead in the ratio of about 84 to 16.

sponding sand-cast bronzes. Provided no appreciable proportions of the brittle eutectoid are present, chill-cast bronzes with their fine grain size would be expected to show a better impact resistance than corresponding sand-cast bronzes with their coarse structure. This is actually the case in the bronzes with 0.25 per cent lead which are considered in Figure 48.

With the addition of appreciable proportions of lead, another factor comes into play. Despite the beneficial effects of small grain size in chill-cast metal, the sand-cast bronzes high in lead showed a higher notch toughness than the chill-cast bronzes of like chemical composition. Lead itself tends to reduce impact resistance, as is shown in Figure 48. The small and finely distributed particles of lead in the chill-cast bronzes (figs. 6 and 7) probably are more effective in reducing impact resistance than the fewer large particles in the corresponding sand-cast bronzes, since small particles more effectively break up the continuity of structure than the same weight of material distributed in large particles.

These features which are associated with grain size are not only a function of rates of cooling but also of pouring temperatures. While the latter variable was under control, detailed records of pouring temperatures were not furnished by the manufacturer. It is therefore possible that differences in pouring temperature were superimposed upon differences in the rates of cooling to give the effects described.

Thus, five important features are shown by the impact tests: (1) Bronzes with small amounts of lead show better resistance to notched-bar impact than the high-lead bronzes. Hence, lead may be said to exert a deleterious effect on the notch toughness of bronzes. (2) Increase of tin in copper within the alpha solid solution range tends to increase impact resistance of bronzes; much more than this amount of tin (which results in the formation of appreciable amounts of the alpha-delta eutectoid) tends to reduce the notch toughness. (3) In the high-lead bronzes containing from about 10 to 25 per cent lead and from 2 to 10 per cent tin the sand-cast metal has a generally better notch toughness than corresponding chill-cast metal. This is probably due to the form and distribution of the lead particles which more than counteract the beneficial effects of the fine-grained matrix in the chill castings. (4) In bronzes with small amounts of lead this order of superiority may be reversed. In such alloys the full benefits are derived from the fine-grained structure, and the chill-cast bronzes may show higher notch toughness than the corresponding sand-cast bronzes with a coarse-grained structure. With intermediate proportions of lead, around 5 per cent, the impact resistance is about the same for the sand-cast and chill-cast metals. Here the benefits of fine grain in the chill castings seem to be balanced by the deleterious effects of the finely disseminated lead. (5) Increase in temperature to 600° F. (315° C.) tends to reduce the differences in impact resistance referred to in items (1) to (4), inclusive, and tends to lower the notch toughness. However, the decrease in impact resistance of the bronzes is more marked between 350° and 600° F. (175° and 315° C.) than between atmospheric temperatures and 350° F. (175° C.), as is shown in Figures 50 and 51.

6. TENSILE TESTS

The results of the tensile tests are summarized in Figures 52 to 58, inclusive. With the exception of the tests at 350° F. (175° C.) made on individual specimens of alloys with different proportions of tin, the charts are based on the averages of the results of two or more tests on each metal in each condition at each temperature.

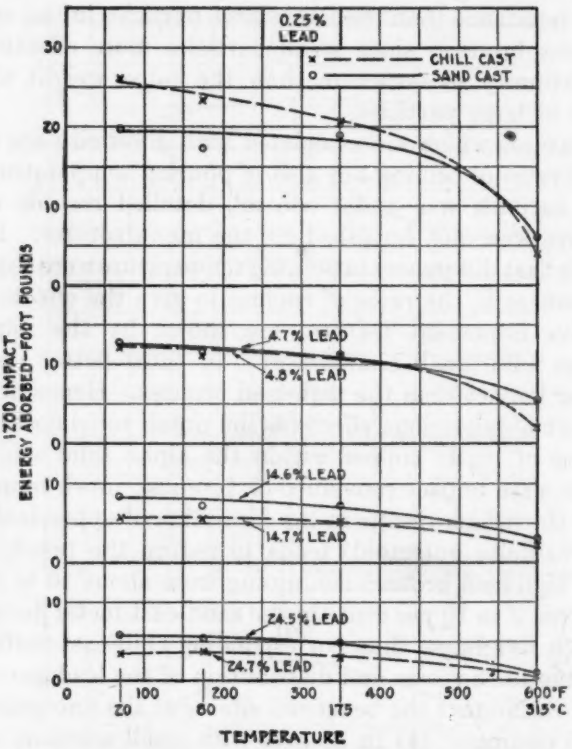

FIG. 50.—*Effect of temperature upon the notched-bar impact resistance of bronzes containing different proportions of lead*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5. These are the same tests referred to in Figure 48.

Increase in lead from 0.25 to around 25 per cent produced a general decrease both in the strength factors and ductility at atmospheric temperatures. On the other hand, increase in tin from a trace to about 10 per cent resulted in a general increase both in the strength factors and ductility. However, as shown in Figure 53, this improvement in tensile properties was much larger for the increase between a trace and 2.5 per cent than between 2.5 and 10 per cent.

In the alloys with different proportions of lead (fig. 52) and likewise those with different proportions of tin (fig. 53) the chill castings

quite generally showed a higher tensile strength than the corresponding sand castings, and this was frequently, but not always, accompanied by decreased ductility.

The values of elongation and reduction of area showed a somewhat wider scatter from smooth curves than the values of tensile strength, due quite largely to the uneven elongation and contraction

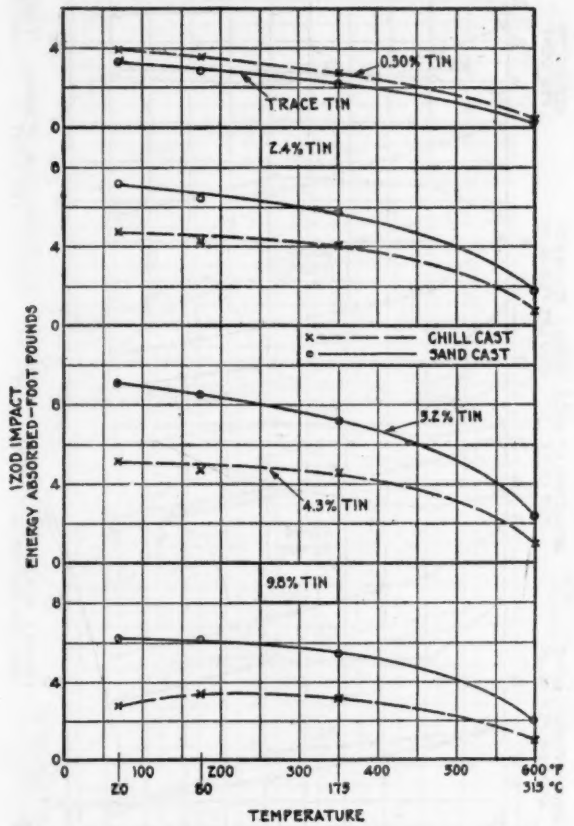

FIG. 51.—*Effect of temperature upon the notched-bar impact resistance of bronzes containing different proportions of tin*

The bronzes contain copper and lead in the ratio of about 84 to 16. These are the same tests referred to in Figure 49.

of these metals in tension tests. This makes it difficult to secure values which adequately represent ductility for general comparisons. However, the tensile tests show that high resistance to static stress is improved in copper-tin-lead alloys by (1) low lead, (2) high tin (up to about 10 per cent), and (3) chill casting. The improvement secured by controlling the first two of these three variables is accompanied by improved ductility; that secured from control of casting conditions may be accompanied either by increased or decreased

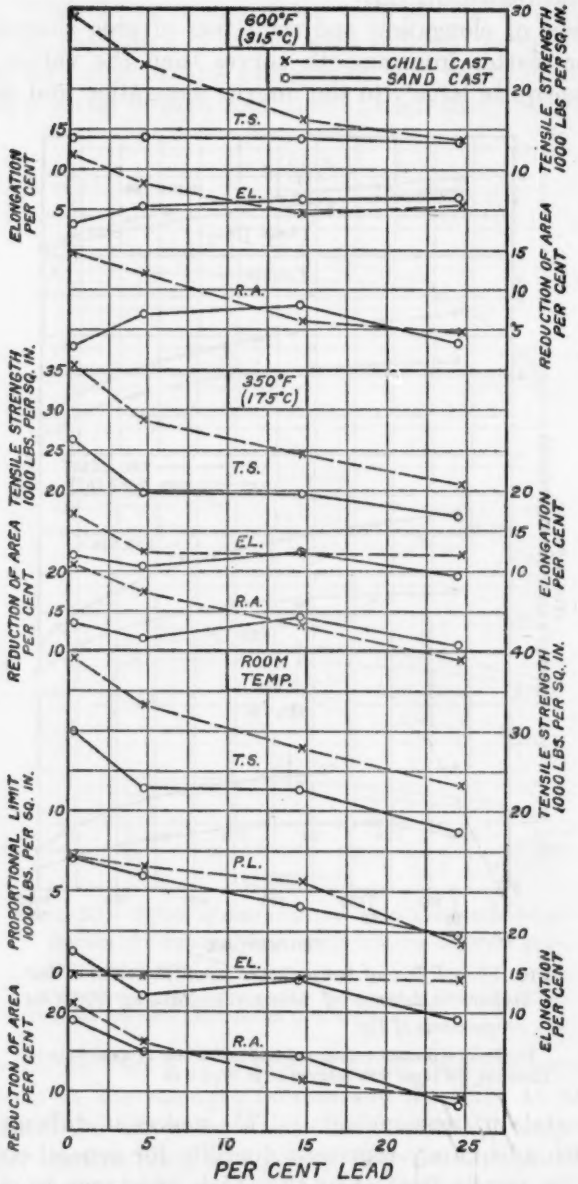

FIG. 52.—*Effect of lead on the tensile properties of bronzes at different temperatures*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5.

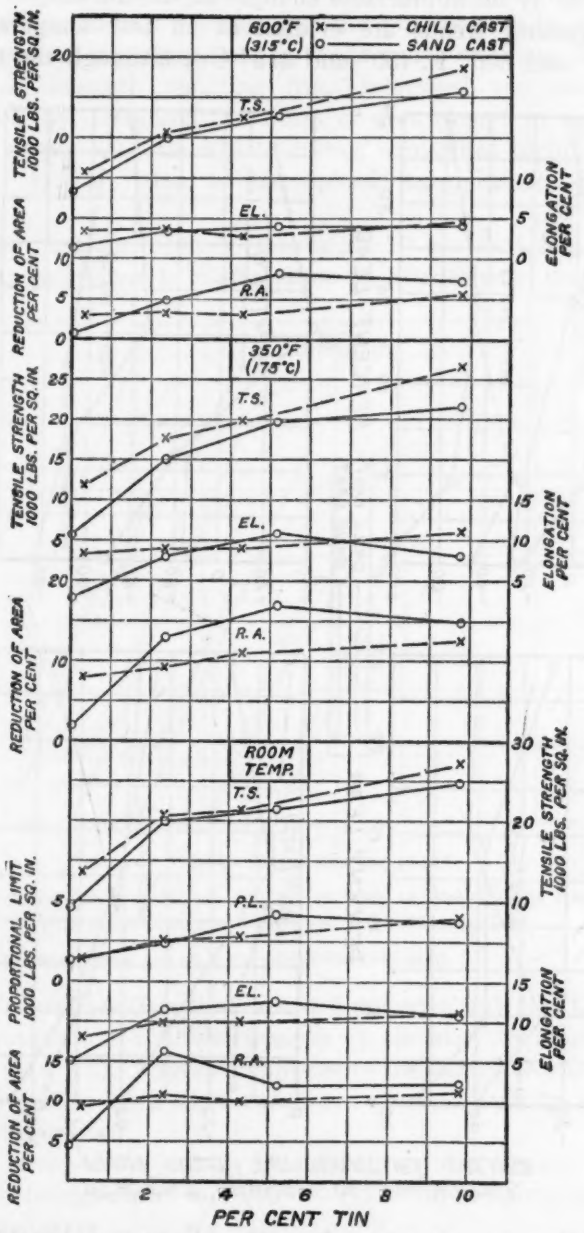

FIG. 53.—*Effect of tin on the tensile properties of bronzes at different temperatures*

The bronzes contain copper and lead in the ratio of about 84 to 16.

elongation and reduction of area in tensile tests at atmospheric temperatures or by no appreciable changes in the ductility.

These general trends are evident at all test temperatures between 70° and 600° F. (20° and 315° C.), although the magnitude

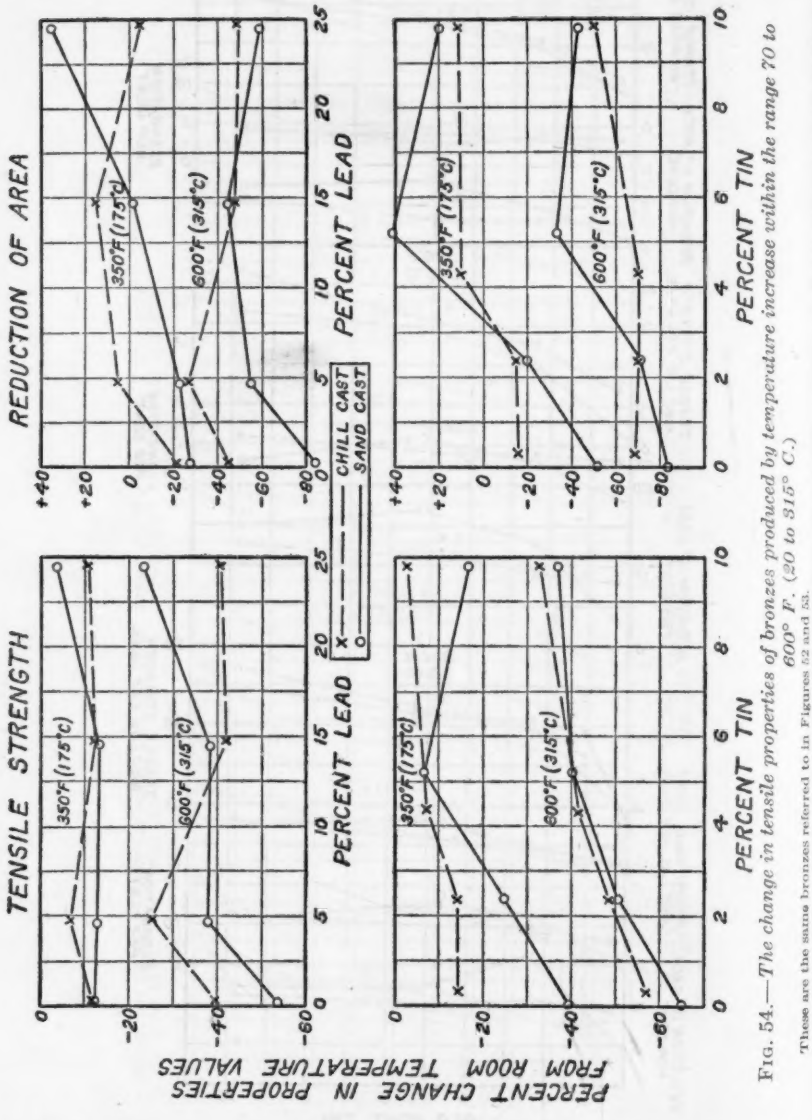

FIG. 54.—*The change in tensile properties of bronzes produced by temperature increase within the range 70 to 600° F. (20 to 315° C.)*

These are the same bronzes referred to in Figures 52 and 53.

of the differences depends upon the alloys and the temperatures considered.

Increase of temperature above atmospheric produced a decrease in the tensile strength of the bronzes (fig. 54), but these changes

became smaller as the tin content of the bronzes increased. They were likewise less in a majority of the chill-cast bronzes than in the corresponding sand-cast bronzes, but did not seem to be affected in any regular manner by variations in the proportions of lead. (Fig. 54.)

Decreased strength resulting from increased temperature was accompanied by decreased reduction of area except in the bronzes with 5 to 10 per cent tin which showed somewhat higher ductility at 350° F. (175° C.) than at atmospheric temperatures or 600° F. (315° C.).

Thus, the high-tin bronzes (maximum 10 per cent tin) not only have higher resistance to static stress at atmospheric temperatures

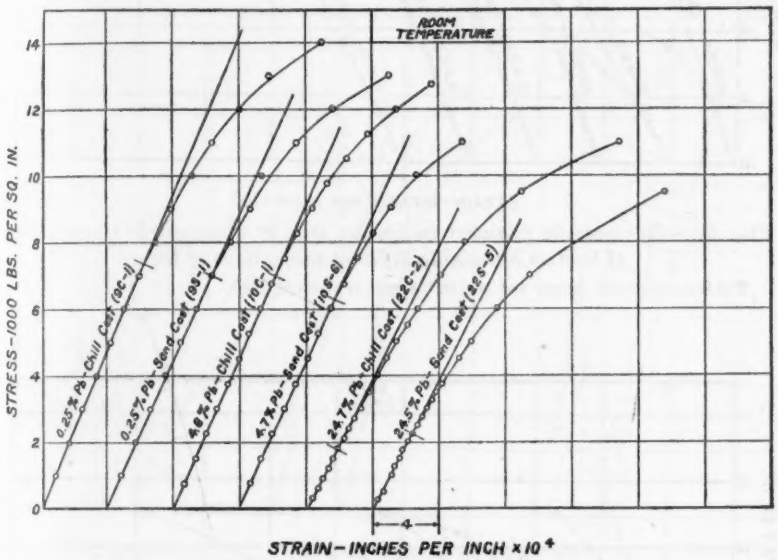
ROOM TEMPERATURE
STRESS-1000 LBS. PER SQ. IN.
STRAIN-INCHES PER INCH x 10⁴
0.25% Pb-Chill Cast (9C-1)
0.25% Pb-Sand Cast (9S-1)
4.8% Pb-Chill Cast (10C-1)
4.7% Pb-Sand Cast (10S-6)
24.7% Pb-Chill Cast (22C-3)
24.5% Pb-Sand Cast (22S-5)

FIG. 55.—*Stress-strain relations in tension tests at atmospheric temperature of bronzes containing different proportions of lead*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5

than low-tin bronzes (less than about 5 per cent tin), but also retain a greater proportion of their strength at elevated temperatures up to 600° F. (315° C.). The same applies to the higher tensile strength of the chill-cast bronzes when compared to sand castings of similar chemical composition.

Typical stress-strain curves secured from the tension tests at atmospheric temperatures and 350° F. (175° C.) are given in Figures 55 to 58, inclusive.

In a number of cases the strain, subsequent to the first applications of stress, decreased with equal increases in the load. This change was sometimes followed by a range in which the stress and strain appeared to be proportional and sometimes by further "steps" in

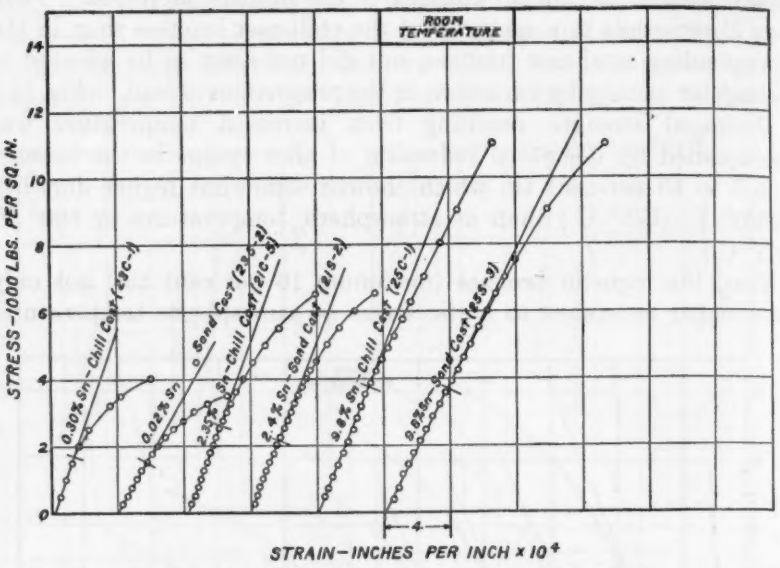

FIG. 56.—*Stress-strain relations in tension tests at atmospheric temperature of bronzes containing different proportions of tin*

The bronzes contain copper and lead in the ratio of about 84 to 16.

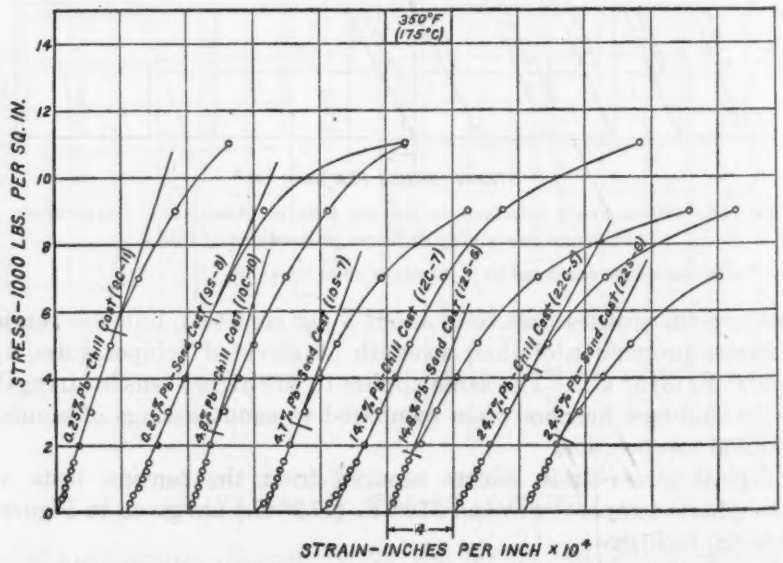
350°F (175°C)
STRESS—1000 LBS. PER SQ. IN.
STRAIN—INCHES PER INCH × 10⁴
0.25% Pb—Chill Cast (9C-11)
0.25% Pb—Sand Cast (9S-8)
4.8% Pb—Chill Cast (10C-10)
4.7% Pb—Sand Cast (10S-7)
14.7% Pb—Chill Cast (12C-7)
14.8% Pb—Sand Cast (12S-6)
24.7% Pb—Chill Cast (22C-5)
24.5% Pb—Sand Cast (22S-6)

FIG. 57.—*Stress-strain relations in tension tests at 350° F. (175° C.) of bronzes containing different proportions of lead*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5.

the stress-strain curves before the strain increased definitely in the expected manner.

These effects were more often observed in tests at 350° F. (175° C.) and were more prominent than in tests at atmospheric temperatures. They also became more marked when the strain was not measured immediately after an increase in the stress but was measured two minutes later.

It is possible that these effects are associated with localized slip at stresses well below the customarily reported proportional limits and the increased importance of the time factor when the temperature is raised from atmospheric to 350° F. (175° C.).

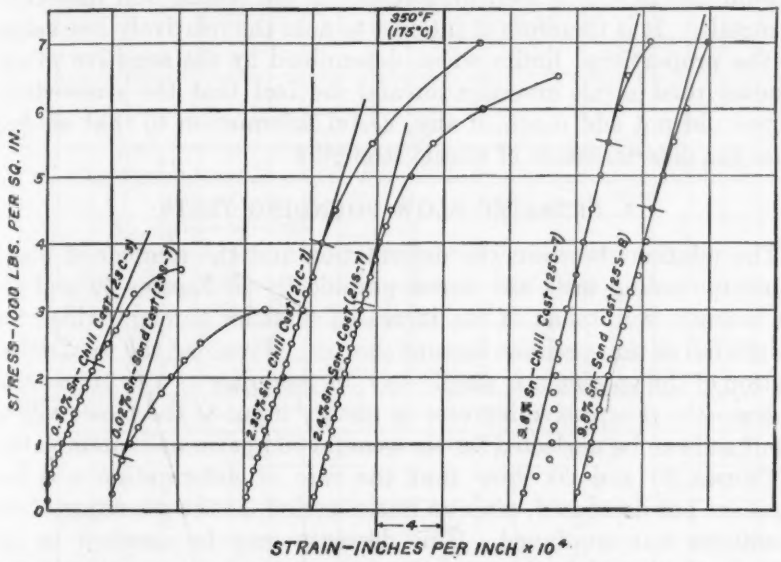
350°F (175°C)
STRESS—1000 LBS. PER SQ. IN.
STRAIN—INCHES PER INCH × 10⁴
0.30% Sn—Chill Cast (25C-1)
0.02% Sn—Sand Cast (25S-1)
2.35% Sn—Chill Cast (21C-1)
2.4% Sn—Sand Cast (21S-1)
3.8% Sn—Chill Cast (25C-1)
8.8% Sn—Sand Cast (25S-8)

FIG. 58.—*Stress-strain relations in tension tests at 350° F. (175° C.) of bronzes containing different proportions of tin*

The bronzes contain copper and lead in the ratio of about 84 to 16.

Handford,[28] with special test equipment, has observed slip at stresses well below the customarily reported proportional limits in open-hearth steel, nickel, cupro-nickel, brass, and copper at atmospheric temperatures. Such slip may be accompanied by more rapid localized strain hardening at 350° F. (175° C.) than at atmospheric temperatures and result in more readily visible discontinuities in the stress-strain curves. In support of this, attention is called to the fact that the proportional limits of some of the bronzes are higher at 350° F. (175° C.) than at atmospheric temperatures.

[28] Handford, C., "A valve method of detecting minute slipping in metals," Phil. Mag., 6th series, 47, p. 806; 1924.

These differences were small, numerically, and it is recognized that the values of proportional limits are not only dependent upon the sensitivity of the test equipment but also upon the methods of interpreting the data. They seem significant in this case, as a majority of the bronzes, especially those cast in chill molds, showed a higher resistance to pounding at 350° F. (175° C.) than at atmospheric temperatures. (See fig. 61.) The pounding tests also emphasized the importance of the time factor at 350° F. (175° C.) in that they also showed evidence of "stepped" deformation in a few cases. (Figs. 59 and 60.)

Bearings are not ordinarily subjected to tension in service, but the tension test is widely used industrially in the testing and inspection of metals. It is therefore of interest to note the relatively low values of the proportional limits when determined by the sensitive extensometer used in this investigation and the fact that the stress-strain curves did not add much, if any, useful information to that secured from the determination of tensile strengths.

7. REPEATED-BLOW POUNDING TESTS

The relations between the deformation and the number of blows in the pounding tests are shown graphically in Figures 59 and 60. No account was taken of the increased distance through which the weight fell as the specimen became shorter. Provided the total deformation of the specimen is small—say, of the order of 5 to 10 per cent or less—the progressive increase in energy input of the blows will be small and can be neglected for the general comparisons contemplated.

Figures 59 and 60 show that the rate of deformation was not constant but decreased, after an initial period, as the pounding of the specimens was continued. This decrease may be ascribed to the work hardening of the metal.

The duration of the initial period of rapid deformation varied considerably with the chemical composition of the bronzes and the method of casting. Likewise, the amount of the first rapid deformation increased with the proportion of lead and decreased with increase in the proportion of tin within the limits investigated. It was also greater, in general, in sand-cast bronzes than in the corresponding chill-cast bronzes.

Similar changes are shown graphically in Figure 61, where comparisons are given of the number of blows required to produce a 5 per cent decrease in the height of the test cylinders. Here the number of blows may be taken to represent the resistance to pounding.

Increase in lead from 0.25 to around 5 per cent produced a marked decrease in the resistance to pounding; with lead increase from around 5 to 25 per cent a further but small decrease was observed.

The resistance to pounding increased slightly with tin increase from a trace to about 5 per cent and showed a very large increase when the tin was raised from 5 to 10 per cent. In practically all cases the

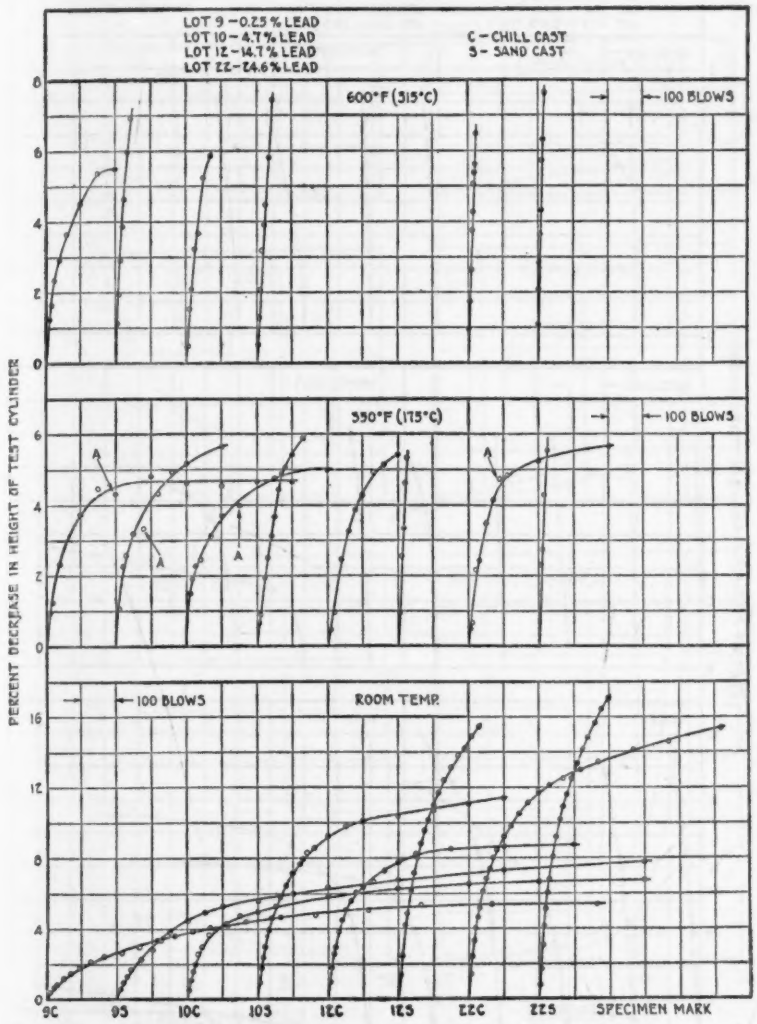

FIG. 59.—*Relation between the number of blows and the deformation in the pounding tests of bronzes containing different proportions of lead*

The bronzes contain copper and tin in the ratio of about 92.5 to 7.5. Letter A and arrows indicate discontinuities in deformation referred to in the text.

chill-cast bronzes showed a higher resistance to pounding than the corresponding sand-cast bronzes.

The described effects from variations in chemical composition and method of casting were similar at 350° and 600° F. (175° and 315° C.) to those at atmospheric temperatures, although the magnitude of

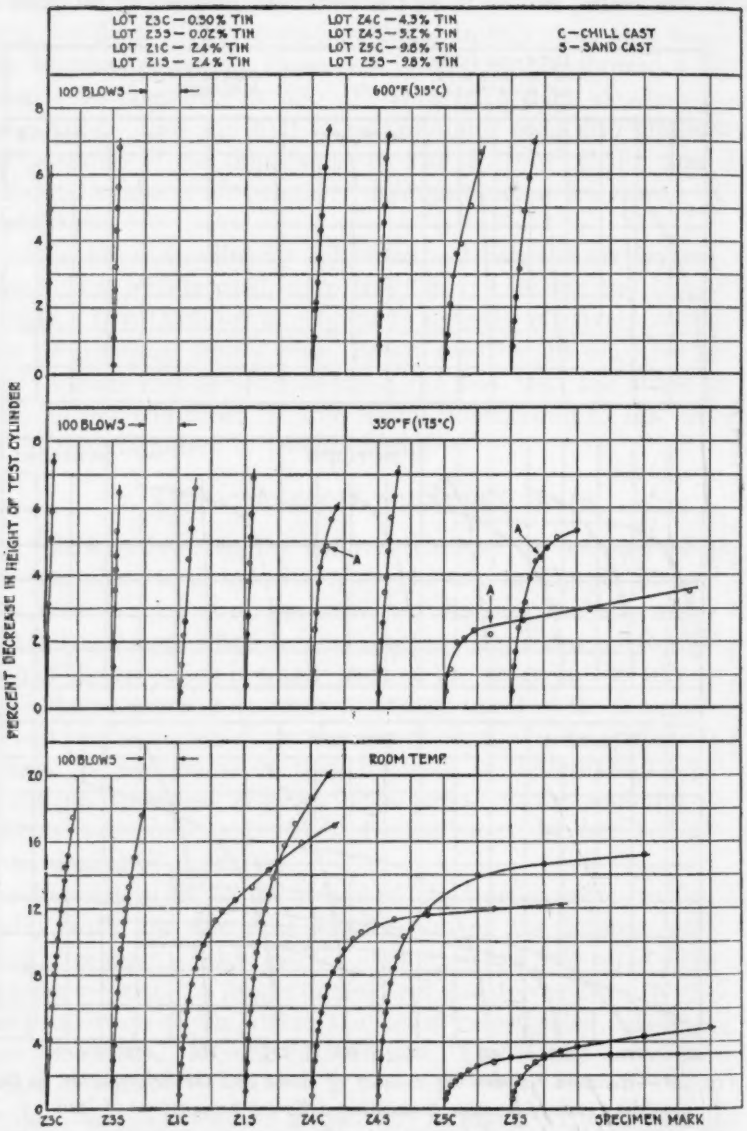

FIG. 60.—*Relation between the number of blows and the deformation in the pounding tests of bronzes containing different proportions of tin*

The bronzes contain copper and lead in the ratio of about 84 to 16. Letter A and arrows indicate discontinuities in deformation referred to in the text.

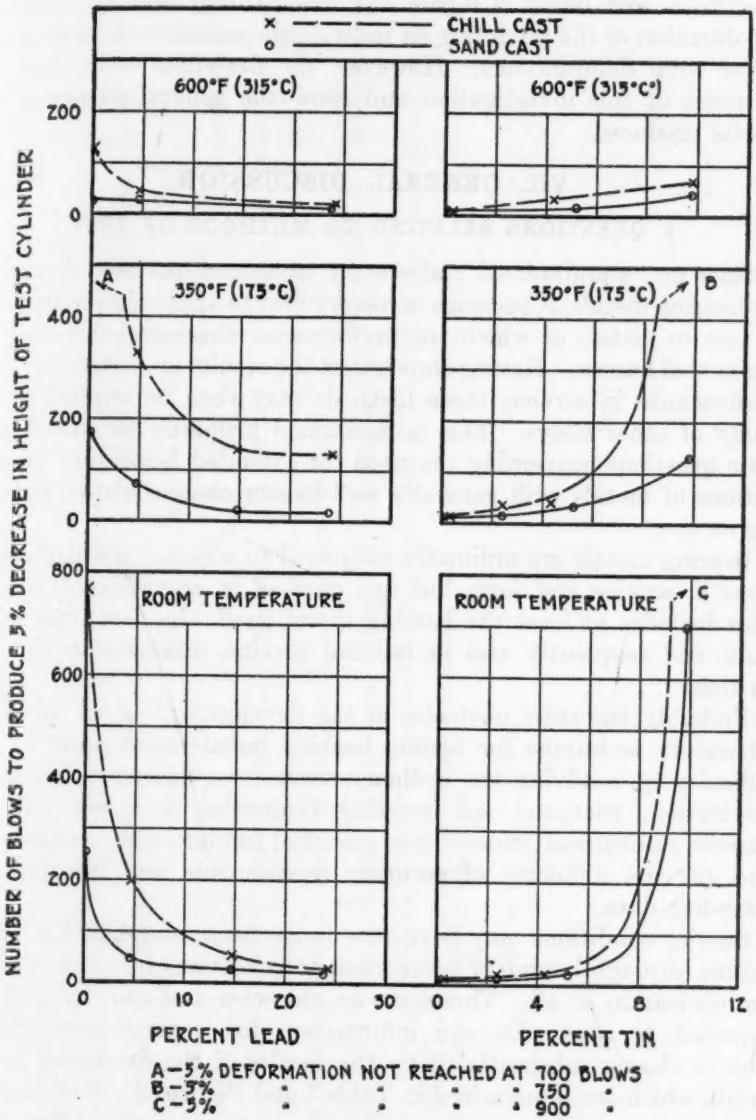

FIG. 61.—*Resistance to pounding of the different bronzes at different temperatures*

These are the same bronzes referred to in Figures 59 and 60.

these effects decreased with increase in temperature, particularly between 350° and 600° F. (175° and 315° C.).

Probably further useful information could have been obtained from tests with blows of higher and lower energy and by increasing the duration of the pounding on individual specimens so as to secure larger total deformations. However, the described tests serve the purposes of this investigation and show the general nature of the effects produced.

VII. GENERAL DISCUSSION

1. QUESTIONS RELATING TO METHODS OF TEST

Since no "standardized" laboratory technique has been developed for bearing metals, it becomes necessary first to apply chosen methods of test to metals of which the performance characteristics are generally well known. Having shown that the results are consistent with performance in service, these methods may then be applied to the study of other alloys. This is mentioned primarily to answer possible questions concerning the need for extended laboratory investigations of metals with generally well-known characteristics in bearing service.

Bearing metals are ordinarily subjected to a complex set of conditions in service and form but one part of a complicated system. This includes at least the bearing metal itself, the shaft, the lubricant, and frequently also in railroad service, nonmetallic abrasive particles.

Probably the chief obstacles in the development of an adequate laboratory technique for testing bearing metals have been (1) the difficulty of resolving the ordinary service requirements, involving lubrication, frictional and wearing properties, and resistance to various mechanical stresses into essential fundamental components, and (2) the difficulty of securing reproducible and interpretable wear-test data.

Service conditions may cover the range from complete film lubrication, through boundary lubrication to conditions in which there is no lubrication at all. Therefore, no one wear test can, by itself, be expected to give adequate information for general comparisons. This is clearly substantiated by the results of the described experiments which are summarized in Table 3 and Figure 62. For example, the wear tests in the presence of oil gave comparisons of the "wearing-in" periods of the bronzes while the tests without oil gave an indication of the behavior which might be expected when lubrication failed in service. Likewise, tests at atmospheric temperatures did not in all cases develop differences corresponding to those at elevated temperatures.

TABLE 3.—*General summary of the results of the wear tests*

Metals containing—	Casting method	Lubricated wear tests [1]		Unlubricated wear tests [1]				Remarks
		Weight loss and duration of "wearing-in" period	Torque after "wearing-in" period	General wear rate [2]	Average torque [3]	Surfaces formed under wear	Corresponding wear on the axle steel	
(3) 84 Cu: 16 Pb; 0.7 per cent tin	Chill			Very high	Low	Rough and flat wheel.	Negligible	Too soft; not a good bearing.
	Sand			do	do	do	do	
(1) 92.5 Cu: 7.5 Sn; 0.25 per cent Pb	Chill	Low	Erratic	High [3]	High	do	Relatively high	Good where good lubrication and low operating temperatures can be maintained.
	Sand	do	do	Intermediate [3]	do	do	do	
(4) 84 Cu: 16 Pb with 5 to 10 per cent Sn.	Chill	High	Low	do	Low to intermediate.		Relative wear intermediate to low.	Good for starved lubrication and elevated temperatures.
(2) 92.5 Cu: 7.5 Sn with 12 to 25 per cent Pb.	Sand	do	do	Low	Intermediate to high.	Strong tendency to film.		

[1] On the Amsler testing machine.
[2] At any temperature between 70 and 350° F. (20 and 175° C.) selected for comparison.
[3] Except at atmospheric temperatures where wear is low.

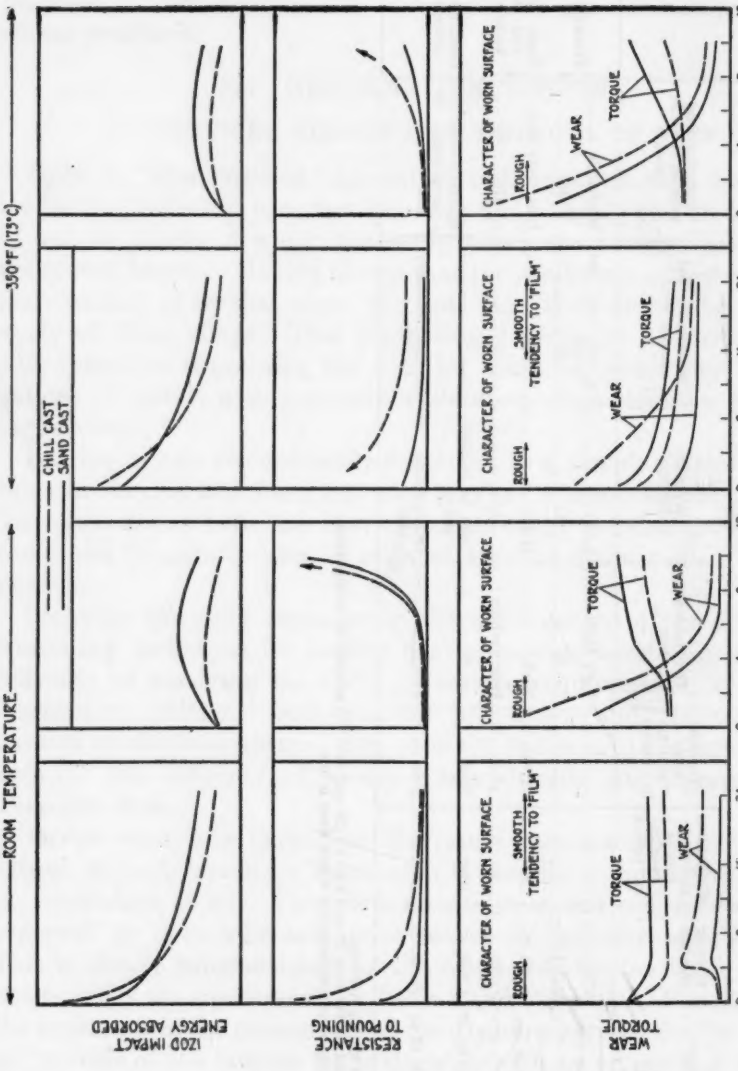

FIG. 62.—*Graphical summary of important features shown in the wear tests and mechanical tests*

Values are not given for the ordinates, as this figure is intended only for general comparisons. Details are included in preceding figures.

Tests of short duration, in which no account is taken of variations in the initial surface conditions of the metals or of possible film formation, are interpreted with difficulty and liable to be misleading. Until such time as definite relations can be developed between the performance in service or corresponding long-time tests and tests of short duration, it seems important to continue tests somewhat beyond the point at which a state of actual or apparent equilibrium is reached.

The wear tests which were made are near two extremes of practical service in that they approach, respectively, conditions of complete film lubrication and the absence of all lubrication. They may not indicate what is to be expected under the intermediate conditions comprising unstable boundary lubrication but gave definitely reproducible data and constitute a foundation upon which to build tests representing other conditions.

It is generally recognized that with complete film lubrication wear is unimportant. It is also probable that difficulties would be encountered in securing reproducible data on wear under conditions of boundary lubrication, so that it seemed logical to turn to unlubricated wear tests which could be carried out without the production of excessive frictional temperatures.

A similar situation is found in considering the results of the mechanical tests as is shown in Table 4. The single-blow impact tests, repeated-blow pounding tests, and tensile tests are not interchangeable, but some seem better adapted than others to the study of bearing bronzes.

TABLE 4.—*General summary of the results of the mechanical tests*

Metals containing—	Casting method	Impact resistance [1] (Izod)	Tensile properties [2] [3]		Resistance to pounding		
			Tensile strength	Reduction of area	70° F. (20° C.)	350° F. (175° C.)	600° F. (315° C.)
84 Cu: 16 Pb; trace to 0.3 per cent Sn	Chill	Low	Low	Low	Low	Low	Low.
	Sand	do	do	do	do	do	Do.
(3) 84 Cu: 16 Pb; trace to 5 per cent Sn	Chill	do	Low to intermediate	do.[4]	do	do	Do.
	Sand	Low to high	do	Intermediate[4]	do	do	Do.
(4) 84 Cu: 16 Pb; 5 to 10 per cent Sn	Chill	Low	Intermediate to high	do.[4]	Low to high	Low to high	Do.
	Sand	High to intermediate	do	do.[4]	do	Low to intermediate	Do.
(1) 92.5 Cu: 7.5 Sn; 0.25 per cent Pb	Chill	High	Very high	High	High	Very high	Medium-low.
	Sand	Intermediate	High	do	Intermediate	Intermediate	Low.
92.5 Cu: 7.5 Sn; 0.25 to 5 per cent Pb	Chill	High to intermediate	Very high to high	High to intermediate [4]	High to intermediate	Very high to high intermediate	Medium-low to low.
	Sand	Intermediate to low	High to intermediate	do	Intermediate to low	Intermediate to low	Low.
(2) 92.5 Cu: 7.5 Sn; 5 to 25 per cent Pb	Chill	Low [5]	do	Intermediate [4]	do	High intermediate to medium low	Do.
	Sand	do	Intermediate	do.[4]	Low	Low	Do.

[1] Increase in temperature above 175° F. (80° C.) caused decrease in impact resistance of bronzes of similar chemical composition.
[2] Increase in temperature caused decrease in tensile properties for bronzes of similar chemical composition. Increase of Pb from 5 to 25 per cent has comparatively little effect on tensile strength of sand-cast bronzes at temperatures between 70 and 600° F. (20 to 315° C.).
[3] Very little elongation change noted with change in chemical composition.
[4] Reduction of area greater at 350° F. (175° C.) than at 70 or 600° F. (20 to 315° C.).
[5] Impact resistance at 350° F. (175° C.) greater than at 70 or 600° F. (20 or 315° C.).

Tension tests, with the sensitive extensometer used in this investigation, require considerable time, particularly when carried out at elevated temperatures, and are therefore expensive to make. They did not show features of importance which were not also developed by the more rapidly made pounding and notched-bar impact tests. Furthermore, the customary tension test specimen is larger than the test specimens used in the two other types of test, and there is therefore a greater tendency toward variations in structure and properties from one part to another in specimens taken from cast metals, particularly from chill castings. While these latter objections can be overcome by using smaller test specimens, the tension test does not as closely simulate important service conditions as do the pounding tests, since bearings are not ordinarily subjected to static tension stresses.

For these reasons the pounding tests and notched-bar impact tests are favored as the first basis of comparison of bearing bronzes. Tensile tests and hardness tests may be added when necessary to meet special requirements. However, there does not appear to be a sharp distinction between the several tests. A more important feature is the interpretation of results.

This is, perhaps, well illustrated by consideration of the bronzes with 5 to 10 per cent tin. Within these limits increase in the tin materially improved the resistance to pounding, but resulted in a slight decrease in the resistence to notched-bar impact.

It will be recalled that within this range of tin content the bronzes changed from the alpha solid solution type to alloys containing the alpha solid solution plus the brittle eutectoid. The presence of this eutectoid evidently exerts a very favorable influence upon the resistance of the bronzes to deformation (pounding), but this is produced at the expense of notch toughness.

The pounding tests can not justly replace the notched-bar impact tests nor are the two to be directly compared. Ordinarily a closer similarity in results would be expected between the tensile strength and the resistance to pounding, but differences will probably also be encountered in such comparisons.

The tests described in this report do not characterize completely the properties which are of interest in the application of bronzes in bearings. For example, the wear tests made at a fixed range of pressures and one rate of slip might be supplemented to advantage by tests made under other pressures and other speeds.

Other properties than those considered are of interest in the application of metals in bearings. One which has not yet been mentioned is the ability of a bearing metal to adsorb a film of the lubricant, so that the film is not readily broken down to permit metal to touch metal (bearing to touch shaft).

Attempts were made [29] with an experimental "oiliness tester" to differentiate among some of the bearing metals, but no useful results were obtained.

2. CHILL CASTING v. SAND CASTING

The sand-cast bronzes quite generally showed lower wear than the corresponding chill-cast bronzes. At first glance this may seem contradictory to the recent experiences of some railroads which have found their over-all costs for locomotive side-rod bearings to be lower with chill-cast bronzes than with sand-cast bronzes. It is not known definitely whether or not these practical comparisons are based on alloys of similar chemical composition, but a sufficient number of such reports have been encountered to justify consideration of the relations between such experiences in service and the described results of laboratory tests.

Under ordinary conditions there is a difference in machining properties in favor of chill-cast metal. Easy machining should not only contribute directly to lower costs, but also indirectly through increased bearing life associated with the better fits which would be a natural consequence of good working properties.

Another matter for consideration is the character of the failure of locomotive side-rod bearings. It has been common practice to state that bearings have worn out when they are no longer serviceable, whether the unserviceability was the result of abrasion, fracture under impact at high temperatures, so-called pounding out, or other causes.

The experience of one railroad has been that "chill brass is largely removed from wear, whereas the sand-cast brass breaks up more frequently in service." [30] This is consistent with the results of the laboratory experiments which showed that the chill-cast bronzes wore more rapidly than corresponding sand-cast bronzes but raises the question of why the sand castings break up more frequently in service since they showed better notch toughness than the chill castings.

Such breakage rarely takes place under the first application of shock in service except in imperfect castings. It is the result of repeated shocks which can be expected to work harden and deform the bronze. As shown in the laboratory pounding tests (fig. 61), the chill-cast bronzes had a higher resistance to deformation than corresponding sand-cast bronzes. Since the sand castings deform more readily, there would be a greater tendency than with chill-cast metal

[29] With a modified Deeley machine devised by W. H. Herschel, associate physicist. The Deeley machine is described in a report of the Lubricants and Lubrication Inquiry Committee of the British Department of Scientific and Industrial Research, 1920. The modifications referred to consist principally of a motor drive and refined temperature control.

[30] Private communication.

for the bearing to "pound out" under the repeated shocks in service. This would have the effect of increasing the shock intensity and possibly account for the fact that the sand-cast bearings more often break up in service than do the chill-cast bearings.

There is, perhaps, room for considerable argument on the question of the superiority of one metal or method of casting over another in respect to the effects of pounding in service. Very few railroad castings are applied without grease grooves of some sort. Service stresses are concentrated at these grooves and result in conditions comparable to those obtaining in the impact tests, although in service the stresses are repeatedly applied as in the pounding tests. Certainly a more complete picture is obtained from the results of both the pounding tests and impact tests than is given by either test alone.

While the foregoing comparisons indicate a closer correlation between some service experiences and the pounding tests, one manufacturer of bearing metals has cited a case where a closer correlation appears to exist between practical performance and the impact tests. His comments [31] are as follows:

> For example, we have the early failures of poured-on lateral plates used for facing the driving truck and trailer boxes adjacent to the hub of the wheel. These lateral plates cause so much trouble by failure from excessive wear and cracking that many of our railroad customers have been replacing the poured-on lateral plate with sand castings applied to the box by means of studs. The poured-on lateral plate is applied at the railroad shops. Molten bronze is poured on the box face after the bronze bearing has been pressed into the box. The lateral is held against the steel box face by means of dovetailing. The lateral plate is then machined and lubrication grooves cut in the brass. The poured-on lateral plate, due to its rapid cooling against the steel face of box, has a very close, fine grain and is very brittle in structure. It is a chill casting in every sense of the word.
>
> Sand-cast lateral plates, however, do not show the failures characteristic of the poured-on lateral plates and are rapidly replacing the old-style plate.

Each application should be given special consideration and tests chosen which will most closely simulate the conditions of service. It is not only necessary to consider the service conditions, but also to consider carefully the interpretation of the results obtained from the selected types of test.

3. EFFECTS OF LEAD AND TIN IN BEARING BRONZES

While the bronzes studied fall within a limited range of chemical composition, there are alloys in the group which should meet a variety of service conditions. Brief consideration will be given to their fields of application as judged from the described laboratory tests, but it should not be inferred that any one alloy is the best for a given purpose, since a narrow range of alloys is considered and subsequent comments are primarily concerned with the interpretation of the laboratory test data.

---

[31] Private communication.

Bronzes with less than about 5 per cent tin do not appear to be suited for bearing service, since their wearing properties and mechanical properties are both relatively poor. (Fig. 62.) They will, therefore, be excluded from further consideration.

There are decided advantages in the high-lead bronzes for service where lubrication can not be maintained. These bronzes containing about 12 to 25 per cent lead did not wear as rapidly as the low-lead bronzes nor assume as rough a surface under friction produced by metal to metal contact without oil. Provided wear is the chief factor to be considered and an approximately constant copper-tin ratio is maintained, variations in lead seem unimportant within the range 12 to 25 per cent. (Fig. 62.)

Where it is also important to have high resistance to pounding, high-notch toughness, and static strength at atmospheric and elevated temperatures, lead may advantageously be kept near the low limit of the specified range, since the strength and notch toughness decrease with increase in the lead. However, these differences are relatively small and for many purposes the bronzes containing from, say, 15 to 25 per cent lead may be considered to be interchangeable.

When the mechanical properties become of primary importance and lubrication can be maintained so that wear is of secondary interest, there are advantages in low lead and high tin. (Fig. 62.) Both favor high strength and resistance to pounding of the bronzes. Low lead and increase in tin up to the point where appreciable proportions of the eutectoid were observed likewise improved the notch toughness. Furthermore, the high-tin bronzes with appreciable proportions of lead and the low-lead bronzes with about 5 to 8 per cent tin maintained a superiority over the high-lead bronzes in their mechanical properties at elevated temperatures, although this was not of the same magnitude as at atmospheric temperatures.

4. SPECIFICATIONS

While variations in the proportions of lead and tin in the bronzes are important in relation to the requirements of service, the effects of some of these changes in chemical composition are much less important than variations arising from the method of casting. For example, the differences in wearing properties produced by increase in lead from 15 to 25 per cent are smaller than those resulting from a change from sand to chill molds. Other examples will be found in discussion of the results of the different tests described in previous sections of this report and in Figure 62.

On the whole, there does not seem to be any real justification for the very wide variation in the specifications of different carriers for bearings of similar design subjected to similar service as outlined in the introduction to this report. It will be seen from Figures 1 and 2 that there are not only wide differences in the type compositions

employed for similar service but also what appear in the light of the described tests to be unnecessary restrictions in the prescribed limits of chemical composition.

The fact that car-journal bearings are replaced without serious difficulties on foreign lines with metals varying widely in chemical composition confirms the viewpoint reached from the laboratory tests that the existing specifications may to advantage be revised and brought more nearly into conformity. Some of the wide variations appear unimportant, and unnecessary restrictions are now encountered. Likewise, a generally improved performance could probably be obtained by more uniform specifications based on the selection of the type compositions best adapted to certain broad requirements.

The effects of impurities upon copper-tin-lead alloys were not considered in this report. Before complete specifications can be drawn information should be obtained on the effects produced by impurities, such as iron and antimony and the effects of elements intentionally added, such, for example, as phosphorus, zinc, nickel, etc. Information on the effect of annealing or other methods of control of the amount of "coring" of the alpha solid solution would also be pertinent.

However, it is of interest to note that the grouping of the bronzes resulting from the described laboratory tests is consistent with the limits of tin and lead in the tentative standards of the American Society for Testing Materials for journal bearings for use on locomotive tenders, passenger and freight equipment cars, specification B67–27T. In these specifications tin may vary from 5 to 7 per cent and lead from 15 to 22 per cent.

VIII. SUMMARY AND CONCLUSIONS

A study was made of the wearing and mechanical properties of two groups of copper-tin-lead alloys in both chill-cast and sand-cast conditions. The first group consisted of bronzes with a copper-tin ratio of about 92.5 to 7.5 and lead varying from 0.25 to 25 per cent; in the second group the bronzes contained a copper-lead ratio of about 84 to 16 and tin varying from a trace to 10 per cent.

The bronzes were tested under rolling and sliding friction without lubricants at atmospheric and elevated temperatures up to 350° F. (175° C.), under lubricated rolling and sliding friction in the presence of oil, and in the presence of abrasives at atmospheric temperatures, and under tension, impact, and pounding at temperatures from 70° to 600° F. (20° to 315° C.). The results of these tests developed the following features:

1. No one of the selected laboratory tests yielded information which was, by itself, adequate for general comparisons. Each of the wear tests and the mechanical tests contributed information of value in developing the characteristics of the different bronzes.

2. Variations in the chemical composition of the bronzes produced major changes in properties, but within certain ranges variations in chemical composition were much less important than variations in the method of casting.

3. As a general rule, chill-cast bronzes wore faster and had lower notch toughness than corresponding sand-cast bronzes, but showed better resistance to pounding and higher tensile strength at temperatures between 70° and 600° F. (20° and 315° C.). The few exceptions to this are shown in detail in the report.

4. In alloys with a practically constant ratio of copper to tin increase in lead produced a general improvement in wearing properties, but this was more marked between 0.25 and 12 per cent than between 12 and 25 per cent lead. At the same time the resistance to pounding, notch toughness, and tensile strength decreased.

5. In alloys with a practically constant ratio of copper to lead increase in tin from 0.7 to around 5 per cent resulted in a marked decrease in the wear produced by rolling and sliding friction without oil and a general improvement in the wearing properties. Further increase in tin from around 5 to 10 per cent did not materially modify the wearing properties but resulted in improved tensile strength and resistance to pounding. There was also an improvement in notch toughness with increase in tin to about 7 per cent, representing the proportion above which appreciable amounts of the brittle alpha-delta eutectoid appeared. With higher tin a slight decrease was observed in the impact values.

6. The bronzes with less than about 5 per cent tin did not seem to have the combination of mechanical and wearing properties necessary for good service as bearings. Where wearing properties are of primary importance the high-lead bronzes have distinct advantages over the low-lead bronzes. Likewise the bronzes with around 8 to 10 per cent tin (and 12 to 15 per cent lead) seem better adapted to service where the chief requirements are resistance to pounding and static stress than bronzes with around 4 to 6 per cent tin.

7. There is a considerable range of chemical composition within which the variations in mechanical and wearing properties are small. This includes a range from 15 to 25 per cent lead and from around 4 to 7 per cent tin. Since alloys within the entire range have given reasonably good performance in locomotive and journal bearings, there does not seem to be justification for the wide variations in specifications now used by different carriers for parts subjected to similar service. This refers not only to the type compositions but also to the limits of chemical composition for any one type.

8. It has been shown that reproducible wear tests can be obtained in the laboratory. The results, when combined with suitable mechanical tests, gave comparisons which were consistent with some experiences in practical service.

ACKNOWLEDGMENTS

In addition to the acknowledgments recorded throughout this report, the authors are indebted to R. F. Staubley, jr., laboratory apprentice, for assistance in many of the wear tests and the mechanical tests; to H. K. Herschman, assistant metallurgist, and S. Epstein, associate metallurgist, for the metallographic work; and to many other of their associates for helpful criticisms and suggestions.

WASHINGTON, May 19, 1928.

STEEL FOR CASEHARDENING—NORMAL AND ABNORMAL STEEL

By S. Epstein and H. S. Rawdon

ABSTRACT

That soft spots in casehardened articles may be due to the kind of steel used, as claimed by McQuaid and Ehn, has been experimentally confirmed. The characteristics of the so-called normal and abnormal microstructures of the extreme types and intermediate gradations are described. On quenching in water structurally abnormal steel proved more prone to give soft spots. However, by quenching in brine or other drastic quenching media uniformly hard cases were obtained in structurally abnormal steel as well as in normal steel. Air or other gases which are usually dissolved in quenching water may cause soft spots. After carburizing, the core of abnormal steel, most probably because of its finer grain, had higher impact strength than normal steel. Most cases of abnormality in commercial steel seem to be associated with the use of aluminum for deoxidizing, although abnormal steel can also be produced in other ways. Nothing arose during the progress of the investigation to disprove Ehn's theory that abnormality is ordinarily chiefly due to the presence of oxides, perhaps dissolved but more probably undissolved.

CONTENTS

	Page
I. Introduction	424
II. Characteristics of the normal and abnormal structure	425
1. Extremes of normality and abnormality	425
2. Intermediate gradations	426
3. Segregations of abnormality	427
4. Depth of carburized layer	427
5. Effect of condition or treatment of the steel	428
(*a*) Mechanical treatment	428
(*b*) Heat treatment	429
(*c*) Annealing in hydrogen	430
III. Quenching tests and hardness measurements for soft spots	431
1. Effect of dissolved gas in water used for quenching	431
2. The structure of soft spots	434
3. Number of soft spots	435
(*a*) Extremes of normality and abnormality	435
(*b*) Intermediate gradations	439
(1) "Fine-grained normal" steel	439
(2) Killed and effervescent steel	439
(*c*) Special manganese carburizing steel	440
IV. Izod impact tests	442
V. Normal and abnormal tool steel	445

Page
VI. Causes of abnormality 447
1. Spectrographic analyses for aluminum 449
2. Deoxidation tests 450
3. Examination of small test ingots taken during progress of open-hearth heats 453
4. Oxygen analyses 453
5. The so-called "iron-carbon-oxygen eutectoid" 456
6. Inclusions 458
7. Thermal analyses 461
VII. Practical aspects 462
VIII. Summary 465

I. INTRODUCTION

In 1922 it was pointed out [1] that soft spots in casehardened articles might be due to the quality of the steel used and not simply to incorrect practice in the carburizing and hardening operations, as might on first thought be assumed. Having noticed in a large roller-bearing plant that soft spots generally occurred only in certain heats of steel, McQuaid and Ehn discovered that these heats showed peculiar microstructural features after carburizing, although otherwise nothing amiss could be detected by the usual chemical or mechanical tests. They became convinced that steel of the peculiar microstructure to which they gave the rather ill-chosen name, "abnormal steel," could not in the ordinary routine of their commercial practice be hardened without soft spots. They also advocated framing specifications whereby steel having an "abnormal" structure in what has become known as the "McQuaid-Ehn carburizing test" would be rejected for casehardening. Ehn ventured the explanation that abnormality in steel is due to oxides, either in solution or as submicroscopic (colloidal) particles.

From the first, considerable skepticism was expressed both as to the conclusion that "abnormal" steel was a cause of soft spots and as to the explanation that "abnormality" was due to oxides in the steel. Most steel manufacturers did not at first believe the steel was to blame for soft spots and objected to including the McQuaid-Ehn test in specifications. On the other hand, many makers of casehardened parts, particularly automobile manufacturers, seemed certain that much of their trouble with soft spots was due to "abnormal" steel. There is still some controversy over the subject, and some metallurgists even yet doubt the existence of abnormal steel. However, in the past few years the McQuaid-Ehn carburizing test has been used widely in steel mills and by carburizers as

[1] E. W. Ehn, "Influence of dissolved oxides on carburizing and hardening qualities of steel," J. Iron and Steel Inst., **105**, p. 157; 1922. W. H. McQuaid and E. W. Ehn, "Effect of quality of steel on case-carburizing results," Trans. Am. Inst. Min. and Met. Engrs., **67**, p. 341; 1922. E. W. Ehn, Irregularities in casehardening work caused by improperly made steel," Trans. Am. Soc. for Steel Treat., **2**, p. 1177, 1922. E. W. Ehn, "Causes of failures in casehardening steel," Iron Age, p. **109**, p. 1807; 1922.

a check on the suitability of the steel for casehardening.[2] The McQuaid-Ehn test has also come into vogue as a supplement to the other metallographic methods. Weber,[3] for example, carburized electric welds made in different atmospheres and found variations in the carburized structures according to the atmosphere used. Wright[4] carburized tool steels to see whether there was any relation between the life of cold-forming tools and the carburized structure of the tools; he found no obvious relation, however.

When requested by several automobile companies to study the "abnormal steel" problem, the Bureau of Standards undertook the study both because of its practical importance and because the fundamental question of quality in steel was involved. It was the aim in this work to determine by laboratory tests whether or not abnormal steel is more prone to give soft spots when hardened after carburization and, if an affirmative answer was obtained, to attempt to ascertain more definitely the cause or causes of abnormality, since an understanding of the causes might allow users to adopt some commercially practicable solution of this phase of the problem of soft spots in casehardening.

The terms "normal" and "abnormal," which have gained currency in this connection, are very inapt; they tend to carry the connotation of good and bad and are also likely to be confused with the term "normalized," which relates to a well-recognized heat treatment. It must be emphasized that by normal or abnormal steel is not meant superior or inferior steel. The terms have been retained for the lack of better ones and are here used in a very restricted technical sense to indicate certain structural characteristics of carburized steel.

II. CHARACTERISTICS OF THE NORMAL AND ABNORMAL STRUCTURE

1. EXTREMES OF NORMALITY AND ABNORMALITY

From McQuaid and Ehn's work, one can hardly dispute that, by carburizing, differences in structure are brought out between steels that otherwise appear to be alike. On the basis of these structural differences, the steel is classed as "normal" or "abnormal." In carrying out the McQuaid-Ehn test the sample is carburized at about 940° C. (1,725° F.) until a thick carburized layer is obtained with a well-defined hypereutectoid zone and is then slowly cooled. Upon examination under the microscope, differences between normal and

[2] W. G. Hildorf, "Improvements in automotive steels," Iron Age, **116**, p. 1378 and 1447; 1925. J. Bethune and W. G. Hildorf, "Gear steels and the production of automobile gears," J. Soc. of Auto. Engrs., **19**, p. 422; 1926.

[3] L. J. Weber, "Studies on electric welding," Trans. Am. Soc. for Steel Treat., **11**, p. 425; 1927.

[4] F. L. Wright, "High-temperature quenching treatment applied to cold-heading ball dies of plain carbon tool steel," Trans. Am. Soc. Steel Treat., **13**, p. 290; 1928.

abnormal steel can be noted in the structure of the core, transition zone, and hypereutectoid zone. The last is usually the most characteristic.

In abnormal steel the grain size is small and the cementite of the hypereutectoid zone is coalesced into thick masses generally surrounded by ferrite. In normal steel the grain size is larger and in the hypereutectoid zone the grains of pearlite are bounded by thin films of cementite. The abnormal structure, therefore, is distinguished by fine-grain size, coalesced cementite, and the presence of a layer of ferrite about the hypereutectoid cementite. These features are generally associated, but not always; some fine-grained specimens show very little coalescence of the cementite, while others of fairly large grain size show pronounced coalescence of the cementite. This will be discussed further under the heading "intermediate gradations." Apparently, the chief feature of the abnormal structure is the lack of perfection in the crystallization of the pearlite, as evidenced by the coalescence of the cementite and its separation from the ferrite. The carburized layer in an abnormal steel is usually thinner than in a normal steel similarly carburized. Figures 1 and 2 show the carburized layers of typical normal and abnormal steel specimens. The characteristics of the normal and abnormal structure are found among the alloy steels as well as in plain carbon steel. Most of the work described in this paper was done on plain carbon carburizing steel of about 0.2 per cent carbon and 0.5 per cent manganese content.

Various kinds of carburizers were tried in the investigation, but the carburizer appeared to play no part in determining the normality or abnormality of the resulting structure. Early in the work a mixture of 60 parts by weight of powdered charcoal and 40 parts barium carbonate was used, but later it was found more convenient to use a commercial carburizer. The carburizing temperature given above must be adhered to fairly closely. If too low a temperature is used, no hypereutectoid zone is produced, while if the carburizing temperature is too high, the structural features of the carburized layer characteristic of abnormal steel become less pronounced, and it becomes difficult to distinguish between the two types.

2. INTERMEDIATE GRADATIONS

Figure 1 illustrates the extremes of structural normality and abnormality in commercial carbon carburizing steel. The differences between the two types are not always so marked, however, there being intermediate gradations, and one is often in doubt as to how to classify a steel according to the McQuaid-Ehn test. Some steel companies have prepared charts showing micrographs of steels classified into as many as 10 grades according to the grain size. These

charts give somewhat uncertain information, for, as Gat [5] has pointed out, classification by grain size alone does not give sufficient weight to the important feature of coalescence of the cementite. In plain carbon steel fineness of grain is usually associated with coalescence of the cementite. With alloy carburizing steels, which are widely used, this is not always so, and a very fine-grained steel may show only slight coalescence of the cementite. Figure 3 shows the structure of the carburized layer of a chromium-vanadium steel (C, 0.50 per cent; Mn, 0.80 per cent; Cr, 0.25 per cent; V, 0.18 per cent) with extremely fine grain, but with only slight coalescence of the cementite. The cementite envelopes in the carburized layer of effervescent steel are usually a little thicker than in killed steel, and because of this effervescent steel may be regarded as less normal than killed steel. However, the extreme types of normal and abnormal steel may be readily differentiated in effervescent steel as well as in killed steel.

3. SEGREGATIONS OF ABNORMALITY

It has been observed that there may be "segregations of abnormality." Figure 4 shows the cross section of a bar of effervescent steel; the inner segregate was abnormal, whereas the outer zone was normal. For a complete McQuaid-Ehn test of a given steel it is necessary, therefore, to inspect a full cross section of a bar and different positions of an ingot. What may be called microsegregations of abnormality are also found, abnormal spots and streaks sometimes being present in samples predominantly normal in structure.

4. DEPTH OF CARBURIZED LAYER

The carburized layer in abnormal steel is usually less deep than in normal steel carburized under the same conditions. In order to obtain a roughly quantitative estimate, microscopic measurements of the depth of the carburized layer were made of a number of samples of normal and abnormal steel, carburized under identical conditions (8 hours at 940° C. (1,725°.F.) in the same pot). The inner boundary of the carburized layer was taken arbitrarily to be at about 0.75 per cent carbon content, and no difficulty was experienced in getting readings showing close agreement. The arrows in Figure 1 show the depth of carburized layer as estimated in this way. As can be seen in Table 1, the average depth of carburized layer of the normal specimens was about 10 per cent greater than that of the abnormal specimens. The effect of varying the time and temperature on the depth of carburized layer of the two types of steel was not determined. It should be noted that the "special manganese carburizing steel" (C, 0.16 per cent; Mn, 1.36 per cent; P, 0.018 per cent; S, 0.096 per cent; Si, 0.07 per cent), this being a sample of a steel now widely

[5] J. D. Gat, "Normality of steel," Trans. Am. Soc. Steel Treat., **12**, p. 376; 1927.

advertised under a trade name for its ease of machining and uniform hardening, had the thickest carburized layer of any of the specimens. Its structure was normal. The properties of this type of steel will be considered in further detail later.

TABLE 1.—*Depth* [1] *of carburized layer of normal and abnormal steel*

Specimen No.	Chemical composition						Depth of carburized layer
	C	Mn	S	P	Si	Ni	
	Per cent	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Thousanths of an inch*
NORMAL							
122	0.14	0.44					42
134	.14	.42					46
146	.12	.43					45
152	.17	.44				0.50	48
55							53
45	.21	.47					55
19M	.12	.37	0.024	0.010	0.063		46
A	.14	.23	.049	.018			50
155	.22	.53	.018	.048	.02		49
Average							48.25
Special Mn carburizing steel J [2]	.16	1.36	.018	.098	.07		59
ABNORMAL							
113	0.14	0.44					42
119	.14	.44					39
125	.14	.42					43
131	.14	.42					44
137	.12	.43					45
143	.12	.43					42
149	.17	.44				0.50	43
56							47
46	.08	.33					41
47	.22	.41					46
13M	.12	.39	0.025	0.012	0.004		42
10M	.12	.34	0.23	.013	.01		42
16T	.13	.37	.027	.010	.008		42
E	.16	.42	.042	.016			46
Average							43.25

[1] See p. 427 for manner of estimating depth.

[2] Material from two different manufacturers was used in the course of the investigation as representative of special manganese carburizing steel. One has been designated "J" and the other "M" throughout this paper.

5. EFFECT OF CONDITION OR TREATMENT OF THE STEEL

(a) MECHANICAL TREATMENT

A question which comes up in classifying a steel by the McQuaid-Ehn test is the effect of the condition or treatment of the steel—whether it is cast, cold-worked, annealed, etc. A few experiments at the Bureau of Standards[6] have indicated that the condition of the steel prior to carburizing generally has little effect on the structure of the carburized layer. A 2-inch round rolled from a 21 by 21 inch normal steel ingot had the same normal structure as the ingot. A one-half inch flat bar cold-rolled from a 1½-inch round normal bar

[6] S. Epstein, Discussion of W. J. Merten: "Irregular carburization of iron and iron alloys—the cause and prevention," Trans. Am. Soc. Steel Treat., **9**, p. 920; 1926.

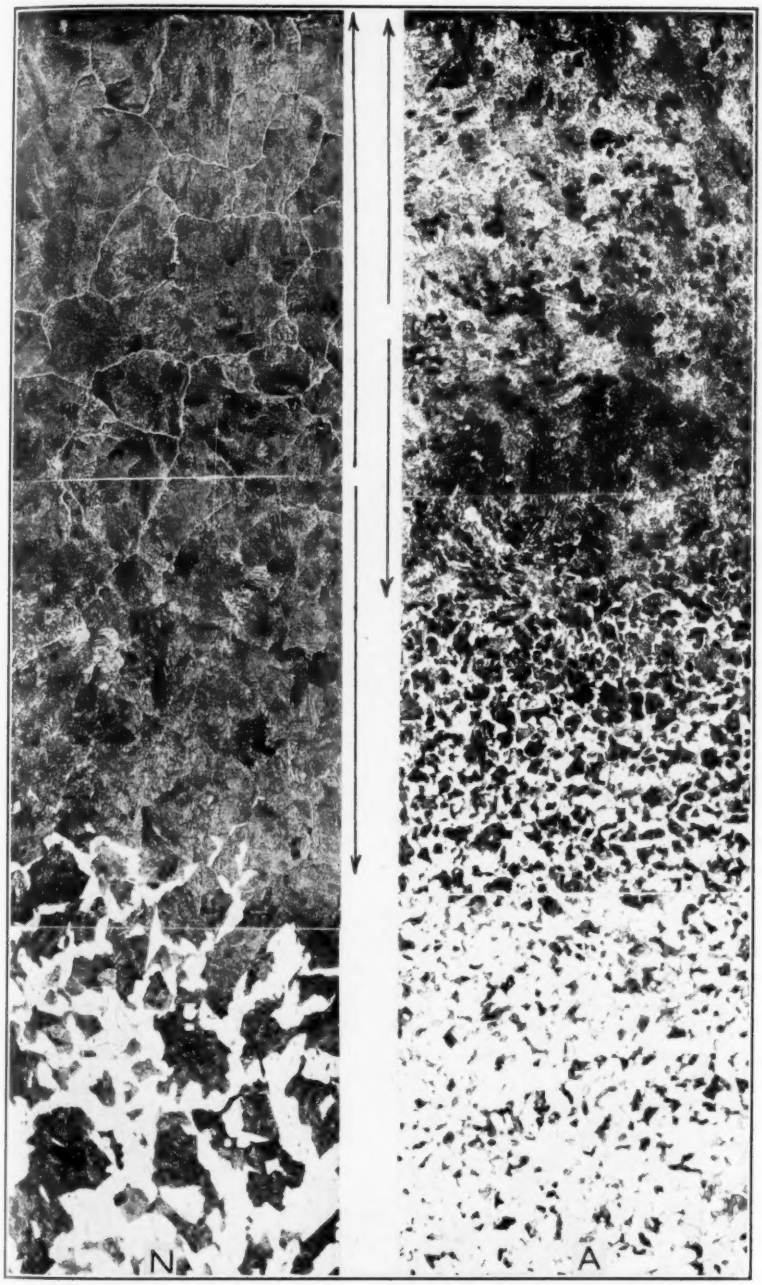

FIG. 1.—*Structure of carburized layers of typical normal and abnormal carburizing steel.* × *100*

The abnormal steel has a finer grain, a shallower case, and shows coalescence of the cementite in the hypereutectoid zone. In all of the micrographs, unless otherwise indicated, the etching reagent was 2 per cent nitric acid in alcohol. The letters, N and A, refer to normal and abnormal, respectively

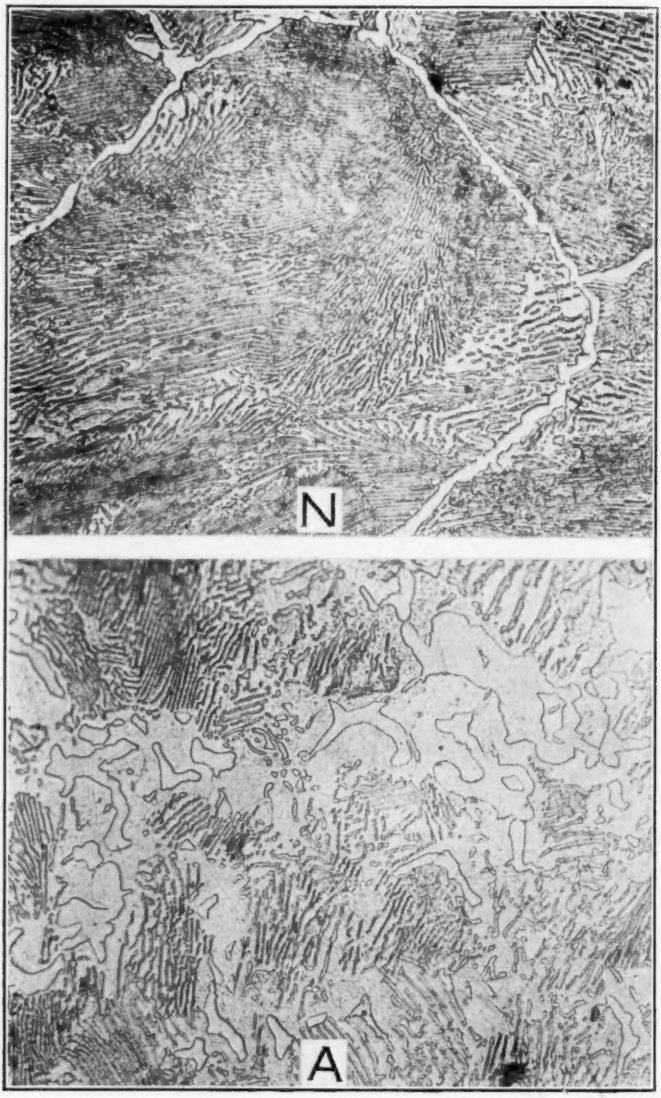

FIG. 2.—*Structure of typical normal and abnormal steel after carburizing.* × *500*

The hypereutectoid zones of the specimens of Figure 1 are shown here at higher magnification. The coalescence of the cementite and its separation from the ferrite in the abnormal steel are very marked. In the normal steel, the pearlite grains are large and well developed, surrounded by thin cementite envelopes

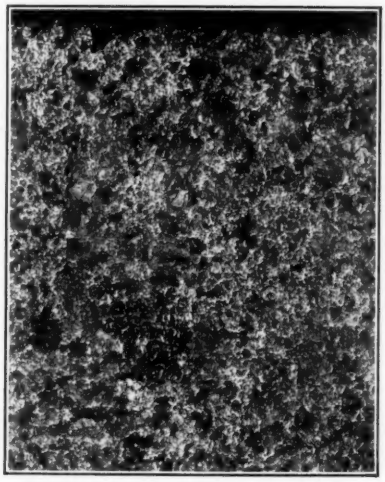

FIG. 3.—*The carburized layer of a chromium-vanadium steel.* × *100*

The grain is very fine, but there is no pronounced coalescence of the cementite

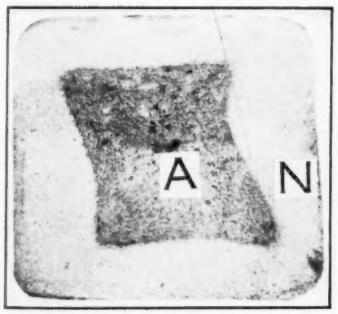

FIG. 4.—*Segregation of "abnormality."* × *1*

A cross section of the bar was deeply etched with hot 1:1 hydrochloric acid. The carburized structure of the inner segregate was abnormal; the outer zone was normal

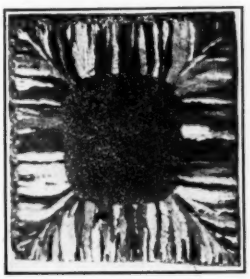

FIG. 6.—*Effect of annealing in hydrogen on abnormal steel.* × 3

Etchant, aqueous ammonium persulphate

Cross section of a specimen of abnormal carburizing steel annealed in hydrogen at 830° C. for 24 hours; the outer coarsely crystalline zone was apparently completely decarburized

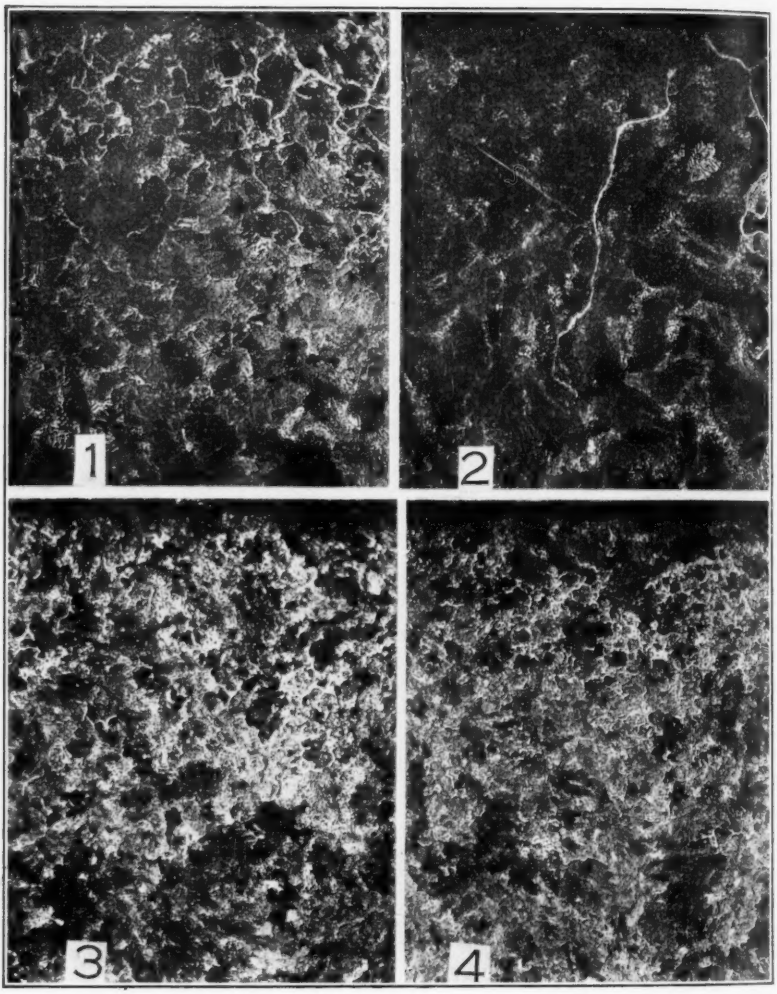

FIG. 5.—*Effect of heat treatment, prior to carburizing, on the structure of abnormal steel.* × *100*

1 and *2*, the carburized layers of a carburizing steel before and after three repeated heatings at 870° C. for 20 minutes and cooling in air. It can be seen that the heat treatment produced a change toward a more normal carburized structure; *3* and *4*, the carburized layers of an abnormal steel before and after three repeated heatings at 870° C. for 20 minutes and cooling in air. The heat treatments had no effect on the carburized structure

had the same normal structure as the original bar. A three-sixteenth-inch thick flat bar cold-rolled from a 1-inch round had the same abnormal structure as the original bar. In neither case did annealing the cold-rolled bar prior to carburizing have any effect on the resulting structure. Merten[7] has stated that a strained condition in steel affects the structure after carburizing and illustrated his point by specimens bent into a U shape, the carburized layers of which were deeper on the outer side of the bend (stressed in tension) than on the inner side (stressed in compression). Tests at the bureau confirmed the difference in thickness of the carburized layers in similar specimens. However, when the carburizing was done on the specimens remaining after cutting away the sides of the U and leaving only the bent portion, no difference in thickness of carburized layer between the outer and inner side of the bend was obtained. This indicated that the shallower carburized layer inside the U was due to a lower concentration of carburizing material in the recess at the bend; that is, the result was due to the shape of the specimen rather than to the compressive strains. The conclusion appears to be warranted that cold-work or strain has practically no effect on normality.

(b) HEAT TREATMENT

In regard to the effect of heat treatment, the Bureau of Mines,[8] which has also been studying the problem of normal and abnormal steel, has reported that in some instances heating an apparently abnormal steel to about 870° C. (1,600° F.) for 20 minutes and cooling in air resulted in a more normal structure on subsequent carburizing. It was also stated that some steels which were not transformed to a more normal structure by a single heating and cooling approached nearer to a normal structure after repeated heating and cooling. These results were confirmed in a general way by tests at the Bureau of Standards. The more pronouncedly abnormal steels, however, were scarcely affected by as many as 10 repeated treatments. Figure 5 shows the structure of a sample of steel which was changed to a more normal structure by three repeated treatments, and of another sample not affected by these treatments.

In the matter of classification, the fact that one steel may be made normal or more nearly so by heat treatment, while another may not be affected, might, perhaps, be taken into account. The process by which the transformation is brought about is of some theoretical interest, since it should throw some light on the cause of abnormality. As a useful means, however, of changing abnormal steel to normal the heat-treatment method appears to be impracticable, since not all steels are affected and most steels would require repeated treatments.

[7] W. J. Merten, Discussion of papers on normal and abnormal steel, Trans. Am. Soc. Steel Treat., 12, p. 416; 1927.

[8] R. B. Norton, Thesis on Abnormal Steel, 1926, Carnegie Institute of Technology.

(c) ANNEALING IN HYDROGEN

A series of tests has recently been described [9] in which normal and abnormal steels were heated or melted in various atmospheres. The statement was made that abnormal steel when melted in vacuo persisted in remaining abnormal in its behavior. At the Bureau of Standards a change in the structure of some steels, from abnormal to normal, has been produced by heating in hydrogen. This effect was confined, however, to the zones decarburized by the hydrogen. Figure 6 shows a cross section of a specimen of abnormal carburizing steel after being heated at 830° C. for 24 hours in an atmosphere of hydrogen. The outer zone has the appearance of having been completely decarburized. The coarse, columnar crystal formation in the carburized zone, similar to that in a casting, is noteworthy. Apparently the crystals of ferrite originated at the surfaces and then grew inward perpendicularly from the surfaces as decarburization proceeded. Austin [10] has reported observations of this effect of hydrogen.

After having been heated in hydrogen, the specimens were sectioned and then carburized, so that a carburized layer was obtained both at the interior, which had not been affected by the hydrogen treatment, and at the surface decarburized by the hydrogen. As can be seen in Figure 7, the interiors of the specimens remained abnormal, but in the portion decarburized by hydrogen the structures had been changed to normal, except in the last specimen, which was originally the most pronouncedly abnormal one. In open-hearth iron, which is generally very abnormal, no change to a more normal structure was produced by heating in hydrogen at 830° C. for 72 hours. The first presumption might be that this effect of hydrogen in converting an abnormal structure to normal is due to the reduction by the hydrogen of oxides present in the steel. Other factors probably enter in, however, since hydrogen is not very active in reducing aluminum oxide. This was present in the first specimen in Figure 7 and apparently was a cause of its being abnormal. The larger-grain size and thus more normal appearance of the carburized structure, after heating in hydrogen, may, to some extent, be a reflection of the coarse, columnar structure formed in the surface layer decarburized by the hydrogen.

[9] O. E. Harder, L. J. Weber, T. E. Jerabek, "Studies on normal and abnormal carburizing steels," Trans. Am. Soc. for Steel Treat., **13**, p. 961; 1928.

[10] C. R. Austin, "Hydrogen decarburization of carbon steels, with considerations on related phenomena," J; Iron and Steel Inst., **105**, p. 93, 1922.

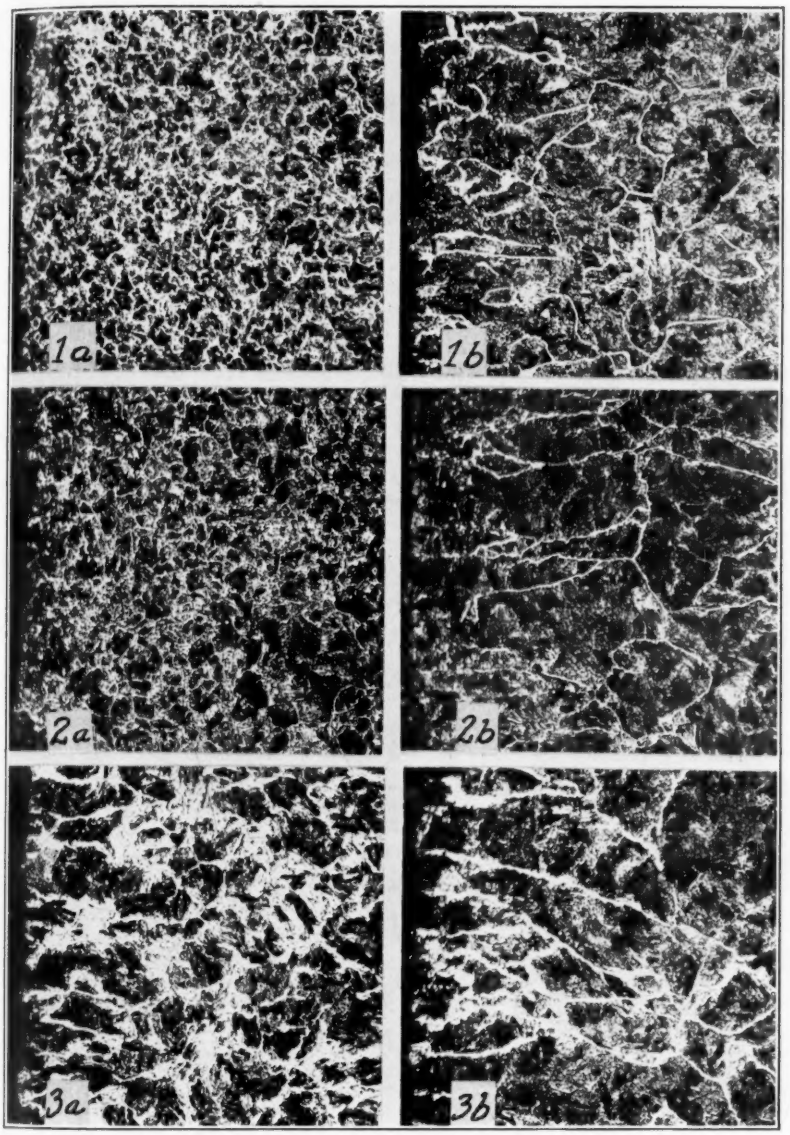

FIG. 7.—*Changes in structure produced in abnormal steel by heating in hydrogen at 830° C. for 24 hours.* × *100*

1a, 2a, 3a, carburized layer in interior not decarburized by hydrogen; the steel remained abnormal; *1b, 2b, 3b,* carburized layer at surface decarburized by hydrogen; the steel appears more normal in structure. Specimens 1 and 3 were of carburizing steel, specimen 2 of tool steel. Specimen 3 did not become so normal as the other two

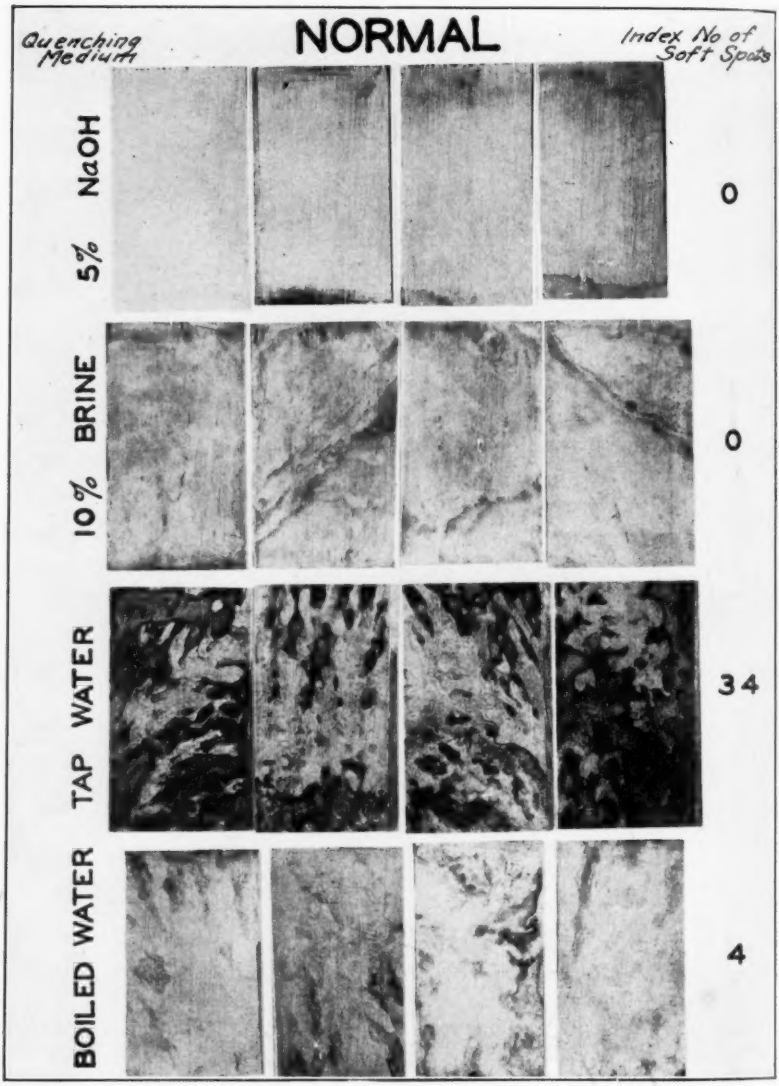

FIG. 8.—*Appearance of the four side faces of blocks of normal steel heated in a cyanide salt bath and quenched in the coolants indicated. About* × ½

The relation is shown between the discolorations on the surfaces and the number of soft spots

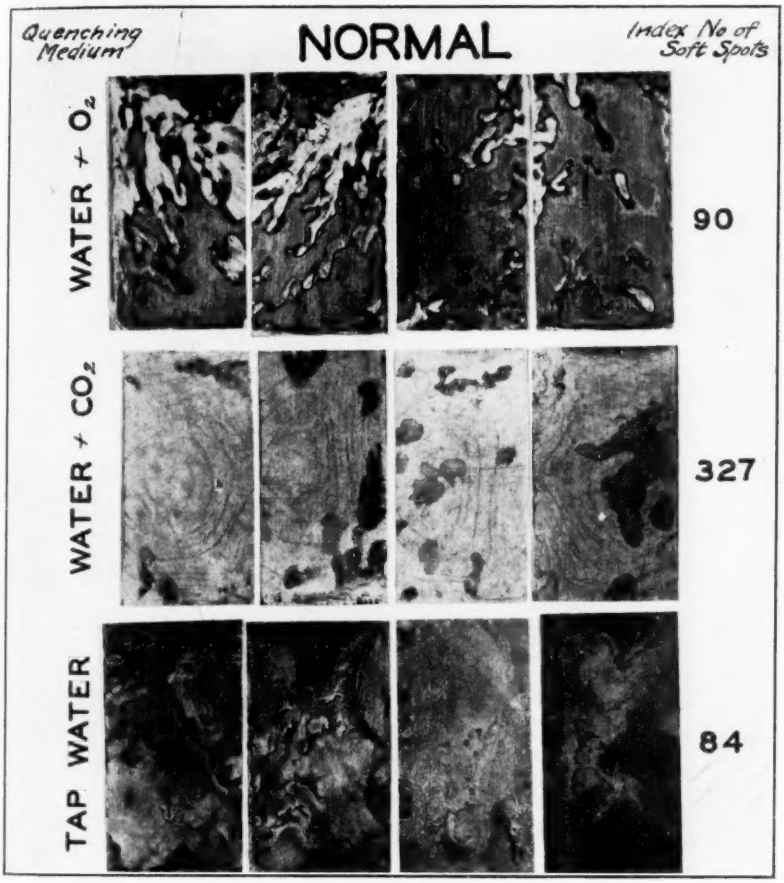

FIG. 8.—(*Continued*)

The last specimen was heated in an electric furnace with access to air, instead of in a cyanide salt bath, before quenching in the tap water. Note the scale on the surfaces

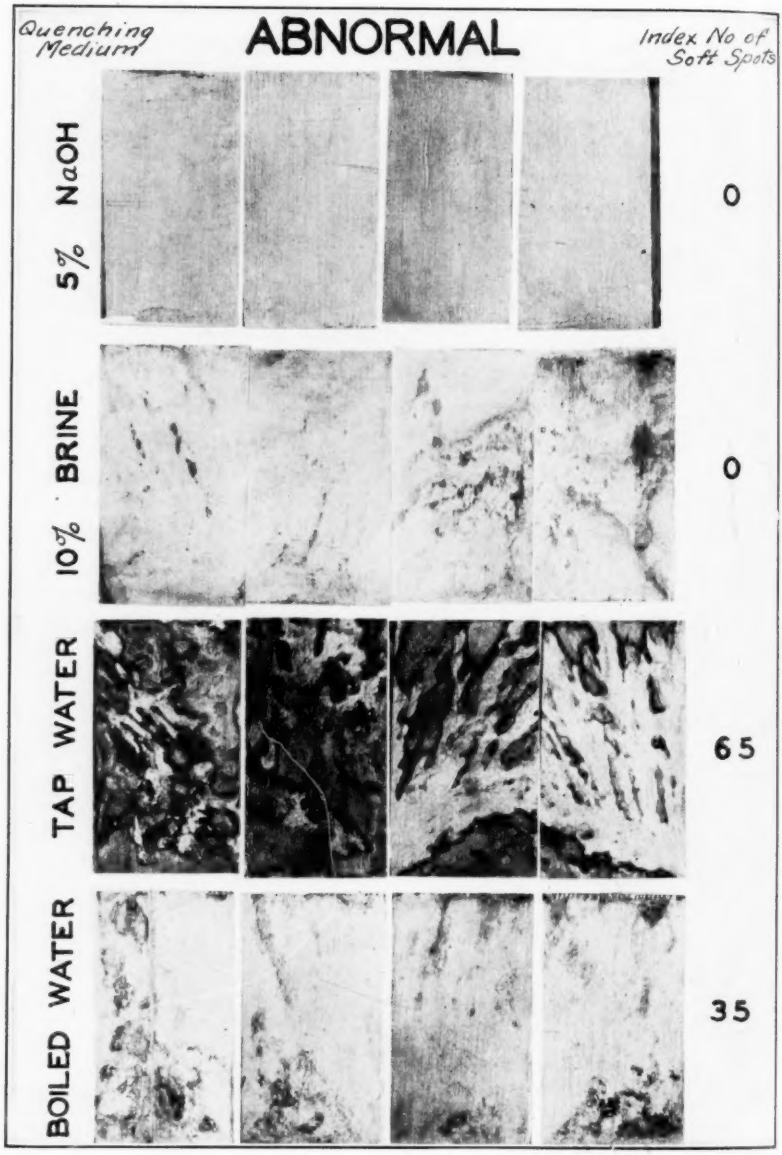

FIG. 9.—*Appearance of the four side faces of blocks of abnormal steel heated as in Figure 8 and quenched as shown. About* × ½

The specimens of Figure 8 were paired with those of Figure 9, during heating and quenching

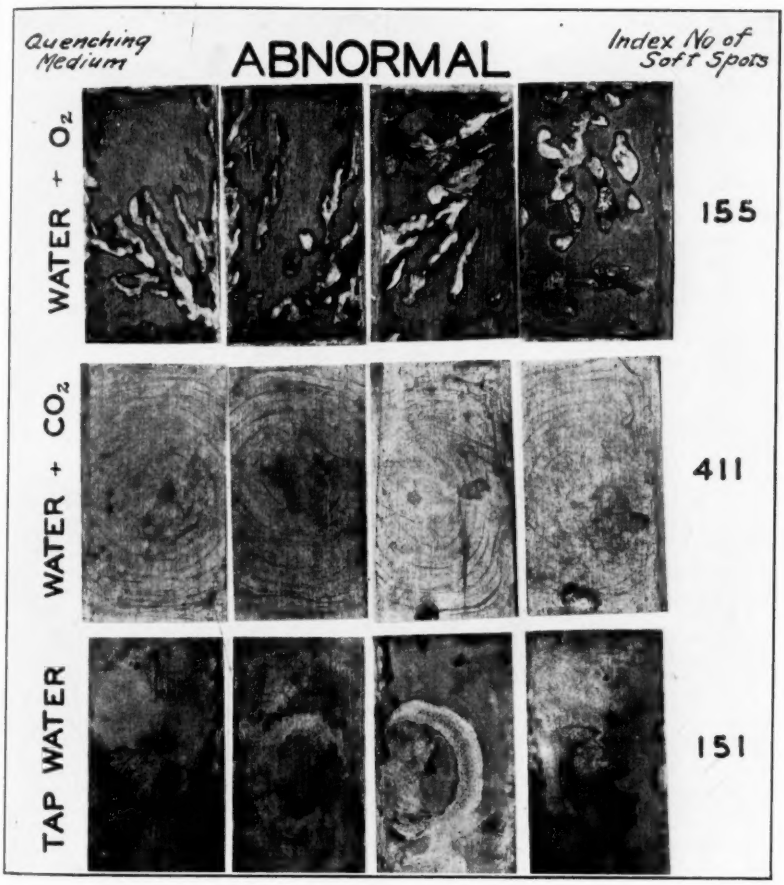

FIG. 9.—(*Continued*)

The last specimen was heated in an electric furnace with access to air, instead of in a cyanide salt bath before quenching in tap water. Note the scale on the surface

UNIV OF MICH

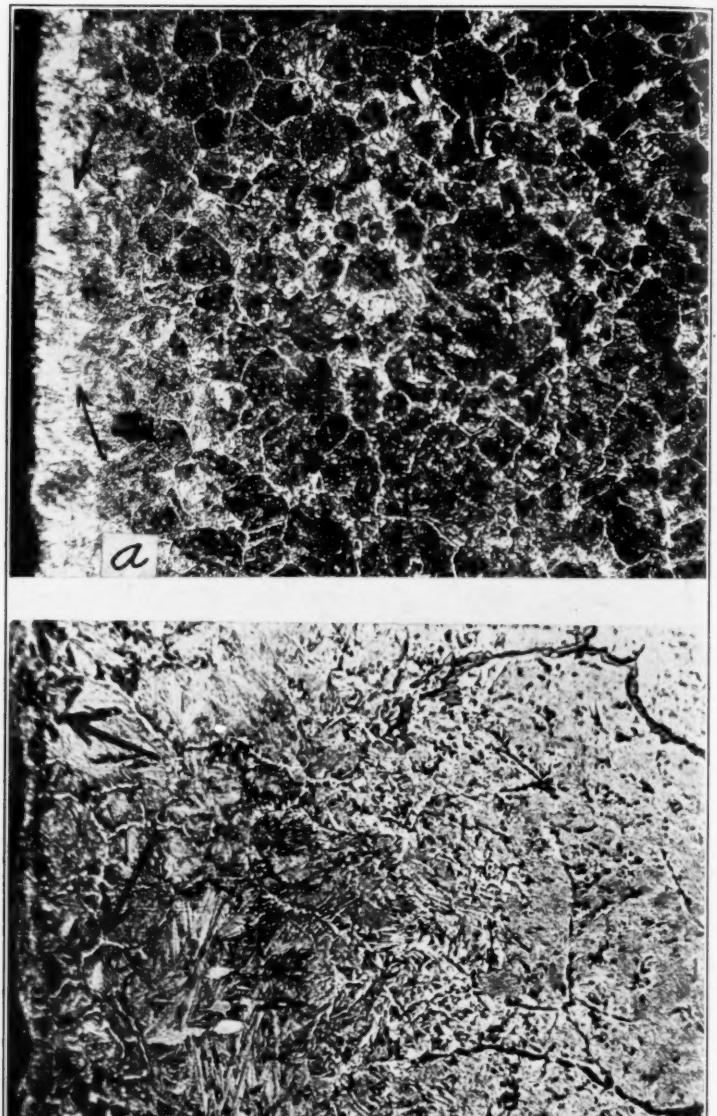

FIG. 10.—*Structure of "file soft" and pseudo "file soft" carburized surfaces.* × *250*

a, carburized layer of 5 per cent nickel steel showing ferrite at surface formed during cooling in the carburizing pot; *b*, carburized layer of normal S. A. E. 1020 steel, heated for two hours in a cyanide bath at 775° C. showing formation of coarsely martensitic layer at surface during prolonged heating in cyanide bath. × 250. Etchant, hot alkaline sodium picrate solution. Note the cracks and hard brittle constituent, probably an iron nitride, at the extreme edge

III. QUENCHING TESTS AND HARDNESS MEASUREMENTS FOR SOFT SPOTS

In the operations of casehardening there are so many possible irregularities which may cause soft spots that the first reaction to the claim that soft spots are to be attributed to the steel was to seek the cause in some shortcoming in the carburizing or quenching operations or, as Hatfield [11] stated it, "indifferent hardening." Much argument about normal and abnormal steel turns largely about this point, and it was the first aim of the present study to show whether abnormal steel is more prone to give soft spots than normal steel. The method used was to quench carburized samples of normal and abnormal steel, of the same size, similarly treated in pairs, and to make Rockwell hardness surveys of the surfaces.

1. EFFECT OF DISSOLVED GAS IN WATER USED FOR QUENCHING

An observation as to the effect of dissolved gas in the water used for quenching should first be described. This behavior of dissolved gas applies to the quenching of steel in general, and not alone to the problem of abnormal steel. In an early series of tests the specimens were heated, prior to quenching, in the usual type of electric muffle furnace, which, of course, is not air-tight. More soft spots appeared in the abnormal than in the normal samples, but the objection [12] was raised to these results that with access to air there is considerable scaling and that the scale on the surface may have been primarily responsible for the soft spots. Heating in a salt bath to which cyanide had been added to obviate scaling eliminated the soft spots on *brine quenching*, but, in repeated tests, soft spots were invariably formed on *water quenching* from the cyanide salt bath.

The soft spots were found to be associated with blue and brown discolorations of the surface. The discolorations appeared to be formed while the specimens were in the water and were finally traced to the presence of dissolved air in the tap water used. When the water was boiled to expel the dissolved gas, the discolorations became much less marked, and there was a decrease in the number of soft spots. When oxygen was bubbled through the boiled water, the discolored areas reappeared, accompanied by an increase in the number of soft spots on specimens quenched in such water. When carbon dioxide gas, which is extremely soluble in water, was bubbled through the water (of course, previous to and not simultaneously with the quenching), no marked discolored areas appeared, since

[11] W. H. Hatfield, Discussion of E. W. Ehn: "Influence of dissolved oxide on the carburizing and hardening qualities of steel," J. Iron & Steel Inst., **105**, p. 198, 1922.

[12] J. W. Napier, Carnegie Steel Co., personal communication.

the carbon dioxide did not oxidize the steel, but practically the entire surfaces of the specimens were soft.

On quenching in 10 per cent brine, in saturated brine, and in a 5 per cent sodium hydroxide solution, the surfaces were not discolored, and there were no soft spots in either the normal or abnormal specimens. This is in agreement with the work of Merten.[13]

Figures 8 and 9 show photographs, about half natural size, of the appearance of the four side surfaces of representative blocks (1½ by 1½ by 2 inches) after these tests. Normal and abnormal blocks, after carburization, were heated in pairs (775° C. (1,425° F.) for 30 minutes) in a cyanide salt bath and quenched in the media indicated. The "index number of soft spots" given for each specimen was obtained empirically in the following manner: 24 Rockwell readings, C scale, spaced equidistantly from each other were taken in three rows on each side of the block, a total of 96 readings for each specimen. In the hard martensitic areas the average reading was about 65. Readings below 60 were taken as an indication of softness. The following weights were given to the "soft readings": From 60 to 55, 1; from 55 to 50, 2; from 50 to 45, 3; below 45, 4. The "index number of soft spots" for any specimen is the sum of the weights of the soft readings out of the 96 readings on the specimen. As can be seen in the photographs, the specimens quenched in sodium hydroxide or brine were not discolored and had no soft spots; those quenched in tap water were discolored and had a good many soft spots; those quenched in boiled water were less discolored and had fewer soft spots; those quenched in boiled water through which oxygen had been bubbled were discolored and had a good many soft spots; those quenched in boiled water through which carbon dioxide had been bubbled were not very much discolored but had a profusion of soft spots. It should be noted that, whenever soft spots were formed, more appeared in the abnormal than in the normal sample, although no difference in the extent of the discoloration was detectable.

The most plausible explanation for the effect of dissolved gas in quenching water is that, when the hot specimen is immersed in the water, the dissolved gas comes out of solution and clings to the specimen, retarding heat abstraction from the interior of the steel because of the low heat conductivity of the gas. In the specimens quenched in tap water and water containing oxygen the soft spots were located in the discolored areas. The discolorations themselves, however, should not be considered as a direct cause of the soft spots; they are simply evidence of the presence of an oxidizing gas. The effect in producing soft spots was similar when an inert gas-like nitrogen, which produced no discolorations, was dissolved in the water.

[13] W. J. Merten, "Fused salt baths for the prevention of soft spots in quenched high-carbon and carburized steels," Trans. Am. Soc. Steel Treat., 7, p. 23, 1925. W. J. Merten, "Irregular carburization of iron and iron alloys—the cause and prevention," Trans. Am. Soc. Steel Treat., 9, p. 907; 1926.

As has been stated, the extent of the discolored areas in the photographs of the surfaces of the quenched specimens is alike in the normal and abnormal samples, but the latter have the larger "index numbers of soft spots." Where gas adhered to an abnormal sample and gave a discolored area, a soft spot was usually formed, the cooling at the spot evidently being slower than the critical cooling rate necessary for the formation of martensite for that type of steel. Normal steel appears to have a lower critical rate, since the slower cooling in the discolored areas was often fast enough to give martensite; fewer discolored areas were soft in the normal steel than in the abnormal steel.

Incidentally, it may be pointed out that quenching tests and hardness surveys so made afford a delicate means of revealing any difference in the critical cooling rate between two steels, it being possible by adjusting the size of specimen and the kind and temperature of the quenching medium to get into the critical range. Slight differences in the critical rate will then throw the balance of the reaction to one side or the other, giving either the Ar′ transformation to troostite, or the Ar″ transformation to martensite, with their measurable differences in hardness. In the same way slight differences in cooling rates of quenching media may be observed, as in the effect of the dissolved air in water, described above. Benedicks,[14] while making direct measurements of the cooling power of water, thought he noticed an effect of dissolved air but finally decided that his measurements showed no perceptible difference between aerated water and distilled water. This result may be called in question, however, since in the same work he also concludes that brine does not surpass water in cooling power, a conclusion which later investigators have disproved.[15]

Water generally contains about 2 per cent by volume of dissolved air (about 1.3 per cent nitrogen and about 0.7 per cent oxygen), depending somewhat on the seasonal changes in temperature. The amount of dissolved air does not differ very greatly for water of different localities. The carbon dioxide content, however, varies considerably, from about 0.2 per cent in mountain-river water to as high as 5 per cent in deep limestone-well water. A large proportion of the dissolved carbon dioxide in deep well water is combined with calcium carbonate as bicarbonate, but the carbon dioxide is readily evolved upon heating, whence the name temporary hard water. In a few rough trial quenching tests made with the hard water of Canton, Ohio, to see if this water, which is reputed to have a high carbon

[14] C. Benedicks, "Cooling power of liquids, quenching velocities, etc.," J. Iron & Steel Inst., 77, p. 153, 1908.

[15] N. B. Pilling and T. D. Lynch, "Cooling properties of technical quenching media," Trans. Am. Inst. Min. & Met. Engrs., 62, p. 2347; 1920. H. J. French and O. Z. Klopsch, "Quenching diagrams for carbon steels in relation to some quenching media for heat treatment," Trans. Am. Soc. Steel Treat., 6, p. 251; 1924.

dioxide content, would produce more soft spots than the Washington water with a low carbon dioxide content, no differences were found. It is well, however, while considering the effect of air dissolved in quenching water, to mention also the possible effect of other gases like carbon dioxide, which may be present in relatively large amounts in the ground water of some localities.

2. THE STRUCTURE OF SOFT SPOTS

Mainly, by way of illustration, it is of interest to examine the structure of soft spots in normal and abnormal steel. Soft spots formed during quenching may readily be recognized by their troostitic structure and thus differentiated from soft spots caused by failure of carbon to penetrate scale on the surface, and from soft areas due to decarburization during cooling in the carburizing pots, which may sometimes occur. The arrow in Figure 10 (*a*) points to a thin ferrite layer at the extreme edge of a carburized specimen of 5 per cent nickel steel, the ferrite layer evidently having formed during cooling in the carburizing pot. The thin surface layer of ferrite was, of course, "file soft" after hardening. Although it may seem paradoxical, apparent file softness of a steel surface may also be produced by prolonged heating in cyanide. During heating in cyanide prior to quenching a nitrogenized layer may be formed at the surface, and, if the heating is prolonged, this layer becomes permeated with cracks. In Figure 10 (*b*), which depicts the structure at the surface of a carburized specimen heated for two hours in a cyanide bath, the outer zone shows a coarse martensitic structure, produced apparently by the penetration of nitrogen. At the extreme outer edge the arrows indicate cracks and fissures and a hard, brittle constituent, probably an iron nitride. The extreme outer layer readily crumbled upon cutting with a file and thus gave the impression of file softness. Such surface embrittlement was not noticed with moderate periods (of about one-half hour) of heating in the cyanide bath.

The appearance of soft spots formed during quenching is illustrated by Figure 11, which shows cross sections of normal and abnormal steel blocks quenched from a cyanide salt bath into tap water, deeply etched with hot 1:1 hydrochloric acid. The difference in grain size between the normal and abnormal steel is noticeable. After deep etching, the troostitic areas appear lighter in color than the martensitic areas, the effect being directly opposite to that produced by etching for microscopic examination. It can be seen that the troostitic areas in the abnormal steel were greater in extent than in the normal steel. Figure 12 shows micrographs of the soft spots. In the abnormal steel the soft spots were more completely troostitic; in the normal steel there was generally some martensite present together with the troostite, especially in the transition zone between the case

and core. The difference in the form of the cementite between the normal and abnormal steel is evident. The white layer at the extreme edge in both micrographs is due to nitride penetration during heating in the cyanide bath prior to quenching.

3. NUMBER OF SOFT SPOTS

(a) EXTREMES OF NORMALITY AND ABNORMALITY

We may now turn to the results of the Rockwell hardness surveys of pairs of carburazed and quenched specimens of normal and abnormal steel, as shown in Table 2. The specimens used were blocks similar to the ones shown in Figures 8 and 9 and had structures corresponding to the extremes of the normal and abnormal structures illustrated in Figure 1. The method of obtaining the "index number of soft spots" has been described above (p. 432). The specimens were heated for 30 minutes at 775° C. (1,425° F.) before quenching. The chemical compositions of the normal and abnormal specimens were practically identical (carbon 0.14 per cent, manganese 0.35 per cent).

TABLE 2.—*Relative tendency of normal and abnormal steel to the formation of surface soft spots upon hardening after carburizing*

[The specimens were heated [1] in pairs and quenched in the designated media]

Pair No.	Heated in—	Quenched		Index number [2] of soft spots	
		In—	At—	Normal	Abnormal
			°C.		
1	Air [3]	Tap water	16	84	200
2	do	do	11	9	11
3	do	do	11	21	86
4	do	do	16	12	90
5	do	do	17	1	21
6	do	do	17	6	45
7	do	do	17	0	96
8	Gas [4]	do	17	135	276
9	do	do	17	87	118
10	do	do	11	1	22
11	do	do	10	17	35
12	Chloride salt bath [5]	do	8	122	193
13	do	do	12	11	131
14	do	do	13	48	168
15	Cyanide salt bath [6]	do	12	43	112
16	do	do	12	52	102
17	do	do	12	4	17
18	do	do	10	1	1
19	do	do	15	0	43
20	do	do	16	19	37
21	do	do	16	17	48
22	do	do	15	24	57
23	do	do	14	17	49
24	do	do	14	26	64
25	do	do	9	28	50

[1] At 775° C. for 30 minutes.
[2] Meaning of index numbers given on p. 432.
[3] Heated in an ordinary electric muffle furnace, which allows access to a limited amount of air.
[4] Heated in an electric muffle furnace filled with illuminating gas.
[5] 78 per cent $BaCl_2$, 22 per cent NaCl salt bath.
[6] A mixture of sodium carbonate and sodium chloride with about 25 per cent sodium cyanide was used. As the cyanide content became exhausted, new additions of sodium cyanide were made to the bath.

TABLE 2.—*Relative tendency of normal and abnormal steel to the formation of surface soft spots upon hardening after carburizing*—Continued

No.	Heated in—	Quenched		Index number[1] of soft spots	
		In—	At—	Normal	Abnormal
			°C.		
26	Cyanide salt bath	Tap water	13	34	65
27	do	do	8	49	173
28	do	do	8	52	149
29	do	do	19	0	10
30	do	do	28	1	133
31	do	do	20	2	14
32	do	do	28	3	20
33	Air	Boiled water	21	66	158
34	do	do	23	14	126
35	do	do	10	12	67
36	do	do	10	12	2
37	Gas	do	11	1	27
38	do	do	12	2	68
39	do	do	24	126	321
40	do	do	25	217	321
41	Chloride salt bath	do	12	14	23
42	Cyanide salt bath	do	8	0	5
43	do	do	9	4	33
44	do	do	3	0	21
45	do	do	3	1	5
46	do	Boiled water plus oxygen	15	90	155
47	do	do	11	36	128
48	do	Boiled water plus hydrogen	8	32	55
49	do	do	8	27	61
50	do	Boiled water plus nitrogen	14	41	165
51	do	do	14	107	144
52	do	do	12	38	146
53	do	do	12	57	184
54	do	Boiled water plus carbon dioxide	16	390	428
55	do	do	17	327	411
56	Air	15 per cent brine[1]	16	0	2
57	do	do	20	0	0
58	Gas	10 per cent brine	15	0	0
59	do	15 per cent brine	22	0	2
60	do	do	26	0	4
61	Chloride salt bath	10 per cent brine	16	0	0
62	do	do	16	0	0
63	do	do	18	0	0
64	do	do	12	0	0
65	do	do	13	0	0
66	do	do	17	0	0
67	do	do	19	0	0
68	do	Saturated brine	18	0	0
69	do	do	18	0	23
70	Cyanide salt bath	10 per cent brine	13	0	0
71	do	do	14	0	0
72	do	do	11	0	0
73	do	do	13	0	0
74	do	do	17	0	0
75	do	Saturated brine	7	0	0
76	do	do	10	0	0
77	do	Saturated brine plus carbon dioxide	10	58	155
78	do	do	10	112	278
79	do	10 per cent brine plus carbon dioxide	8	167	314
80	do	do	8	145	274
81	Air	5 per cent sodium hydroxide	24	0	0
82	do	do	22	0	0
83	Gas	do	23	0	0
84	do	do	20	0	0
85	do	do	21	0	0
86	do	do	22	0	4
87	Chloride salt bath	do	12	0	0
88	Cyanide salt bath	do	8	0	0

[1] Sodium chloride brine.

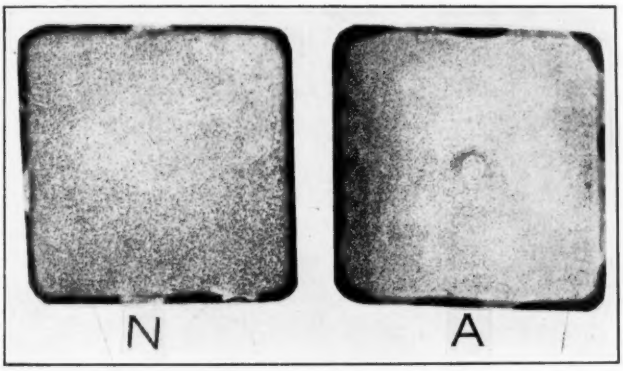

FIG. 11.—*Cross sections of normal and abnormal steel blocks heated together in a cyanide bath and quenched simultaneously in tap water; deeply etched with hot 1 : 1 hydrochloric acid.* × *1*

In the carburized layers, the dark portions are hard, and the light portions are soft spots. It can be seen that there are larger soft areas in the abnormal specimen

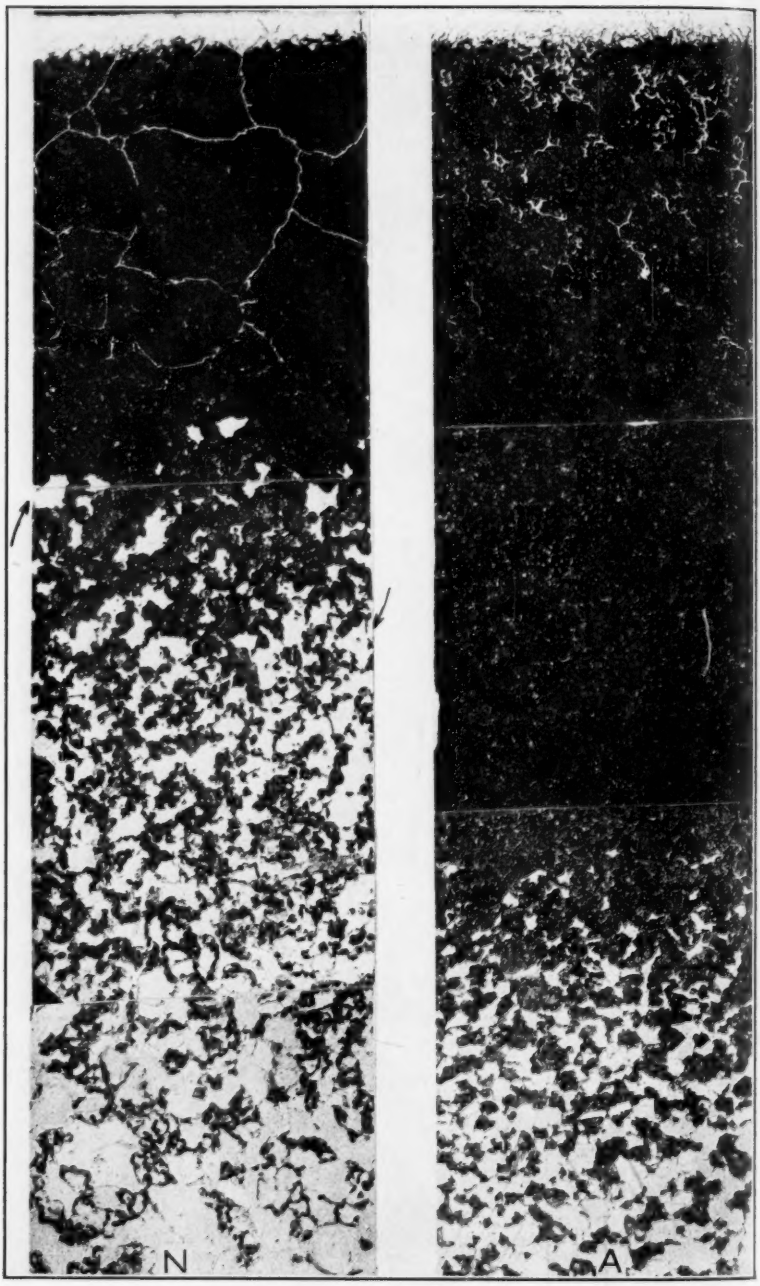

FIG. 12.—*Structure of soft spots in normal and abnormal casehardened steel.* × *100*

In the carburized layer of the abnormal steel the soft spot is completely troostitic, whereas in the normal steel there is some martensite present, together with the troostite. The martensite is indicated by arrows. The white border at the extreme outer surface in both specimens is due to nitrogen penetration during the heating in the cyanide bath prior to quenching

The data in Table 2 show a consistent difference in the number of soft spots between normal and abnormal steel and clearly indicate that abnormal steel is more prone to give soft spots. The data of Table 2 also show that with the use of a drastic quenching medium, such as sodium chloride brine or sodium hydroxide solution, the formation of soft spots can be completely prevented in both normal and abnormal steel. A word may be said as to the effect of the heating medium. Although in some of the tests more soft spots were obtained upon heating in an air or gas atmosphere, where scaling was possible, than upon heating in a cyanide salt bath, the data in Table 2 indicate that, as regards the elimination of soft spots, the type of quenching medium used is of much greater importance than the manner of heating.

The advisability of changing from water quenching to a more drastic medium like brine, water-pressure spray, or sodium hydroxide solution, deserves careful consideration. With most casehardened articles there is small danger from cracking. The question is whether drastic quenching will increase distortion or warpage. The view might be taken that warpage is, perhaps, more a result of uneven-temperature gradients than steep gradients, and that there would thus be no increase in troubles from warpage if the more drastic methods of quenching were adopted to avoid soft spots. However, it would require considerable study to ascertain this point. To throw some possible light on it, a few trial quenching tests were made of carburized hollow cylinders, 2½ inches long, 1½ inches outside diameter, and ¼-inch wall thickness. Very precise measurements [16] were made, with a Prestwich gauge, of the lengths and diameters of the cylinders before and after quenching in oil, water, brine, and sodium hydroxide solution. Two types of steel were used, normal S. A. E. 1020 steel (C, 0.22 per cent; Mn, 0.53 per cent; P, 0.018 per cent; S, 0.048 per cent; Si, 0.02 per cent), and the special manganese carburizing steel M (C, 0.19 per cent; Mn, 1.04 per cent; P, 0.060 per cent; S, 0.062 per cent; Si, 0.13 per cent). Three elements of length at angles of 120° were measured, and three measurements were made on each specimen of the diameters, at the two ends and middle of the cylinder along each of the three elements, before and after quenching. A survey was also made of the number of soft spots in the quenched specimens. The specimens were then tempered for one hour at 200° C. (392° F.) and similar measurements made. The arithmetical sums of the maximum positive and negative deviations from the mean change in dimension from the original size of the carburized cylinders, after quenching and after tempering, is given in Table 3.

[16] Ralph F. Staubley, jr., laboratory aid, Bureau of Standards.

TABLE 3.—*Dimensional changes produced in carburized hollow cylinders by quenching in various media*

[Arithmetical sums of maximum positive and negative deviations from mean change in dimension, in thousandths of an inch]

Quenching medium	Special manganese carburizing steel [1]				S. A. E. 1020 steel [1]			
	Specimen No.	Length		Diameter	Specimen No.	Length		Diameter
		Quenched	Tempered	Quenched / Tempered		Quenched	Tempered	Quenched / Tempered
Oil	A-1	0.22	0.18	0.38 / 0.66	B-1	0.45	0.57	2.37 / 2.57
	A-2	.04	.14	.56 / .96	B-2	.13	.21	1.17 / 1.19
	A-3	.37	.55	.57 / .65	B-3	.37	.13	2.63 / 2.43
		(No soft spots)				(Very many soft spots)		
Water	A-4	.42	.43	1.55 / 2.00	B-4	.57	.50	2.95 / 2.70
	A-5	.18	.09	1.47 / 1.52	B-5	.27	.31	1.18 / 1.63
	A-6	.47	.34	1.38 / 1.38	B-6	.26	.23	1.43 / 1.38
		(No soft spots)				(Soft spots only in specimen B-4)		
5 per cent NaOH	A-7	.57	.29	3.25 / 3.25	B-7	1.20	1.20	2.40 / 1.90
	A-8	.73	1.05	1.74 / 1.84	B-8	.78	.18	2.86 / 2.43
	A-9	.65	.75	1.75 / 2.65	B-9	1.03	1.28	2.75 / 3.23
		(No soft spots)				(No soft spots)		
10 per cent brine	A-10	.66	.86	2.88 / 2.70	B-10	1.04	2.07	2.25 / 2.15
	A-11	.77	.80	1.92 / 1.60	B-11	.27	.38	1.17 / 2.48
	A-12	1.77	1.87	1.97 / 2.69	B-12	1.44	.59	2.43 / 2.48
		(No soft spots)				(No soft spots)		

[1] See p. 437 for chemical composition.

It seems justifiable to take the deviations from the mean change in dimensions in Table 3 as an indication of the amount of warpage. In the special manganese carburizing steel, which, it will be noted, hardened without soft spots even on oil quenching, the least "warpage" occurred on oil quenching. Upon water quenching there was more "warpage," and there was still more with the more drastic quenching media—brine and sodium hydroxide. In the plain carbon steel, which, it will be noted, did not harden in oil, there was about the same amount of warpage on oil quenching as on quenching in the more drastic media. The large amount "warpage" of the plain carbon steel upon oil quenching may have been due to the nonuniform hardening. In accord with this assumption, specimen B–4, the water-quenched plain carbon steel which had some soft spots, whereas its two companion specimens did not, showed the most "warpage" in its group. The data at hand are much too meager for definite conclusions, but they seem to indicate that the least warpage occurs with the use of the mildest quenching medium that will still give uniform hardening. For sizes in which the special manganese steel or similar steel would harden uniformly on oil quenching, whereas plain carbon steel would not, this fact would appear to make it possible to obtain less warpage with such alloy steels than with plain carbon steel.

The more drastic quenching media, such as brine, are being very generally used, however, especially for large casehardened articles.

High-pressure water-spray quenching has also found favor. An advantage of quenching in brine and especially in sodium hydroxide is the very clean bright surface of the work obtained.

(b) INTERMEDIATE GRADATIONS

(1) FINE-GRAINED NORMAL STEEL.—As has been stated, the difference in the number of soft spots between normal and abnormal steel shown in Table 2 was obtained with steels showing the extremes of the normal and abnormal structure. No definite data are available for the intermediate gradations. Among these, interest centers mainly on the fine-grained steels showing no coalescence of the cementite. To determine the tendency to form soft spots of such steels would necessitate making quenching tests and hardness surveys of specimens of the same composition, showing a wide variation in grain size and no coalescence of the cementite. Such specimens could not be procured, however, the reason being that in the plain carbon steels small-grain size was always accompanied by more or less coalescence of the cementite. In some alloy steels, on the other hand, it was possible to get fine-grain size without, or with only very slight, coalescence of the cementite, but these alloy steels were always fine grained, and no coarse-grained material of the same composition could be obtained for comparison. Gat [17] has presented data intended to show that in a steel with no coalescence of the cementite fine-grain size does not indicate any tendency to form soft spots. However, he used only the fine-grained alloy steels, without making any comparison tests with similar coarse-grained steels, and it is most probable that the uniform hardness he obtained was due to the deep hardening properties conferred by the alloying elements. Whether or not, therefore, fine-grained size of itself, and when not accompanied by coalescence of the cementite, should be regarded as a feature of abnormality and as an indication that a faster quenching rate is necessary for uniform hardening than with coarse-grained steels of the same composition remains unsettled. However, the question here involved is academic rather than practical, since the fine-grained alloy steels do not appear to be prone to form soft spots in practice. These steels have been classed by some metallurgists as "fine-grained normal" and are regarded as particularly desirable, it being considered that their fine grain allows the so-called "core-refining heat" or "regenerative quench" in the heat treatment after carburizing to be dispensed with.[18]

(2) KILLED AND EFFERVESCENT STEEL.—Another intermediate grade is that of a coarse-grained steel showing somewhat thickened cementite envelopes, with occasional thin ferrite films about the

[17] See footnote 5, p. 427.

[18] M. T. Lothrop, Casehardening, Iron Age, **120**, p. 612; 1927.

cementite. This structure is usually found in effervescent steel, and the conclusion is sometimes drawn that the distinction between normal and abnormal steel reduces itself in the main to one between killed and effervescent steel. This conclusion was not borne out by the quenching tests and hardness surveys shown in Table 4. Coarse-grained effervescent steel with thick cementite envelopes gave only a few more soft spots than normal killed steel. The difference between normal killed steel and normal effervescent steel was negligible, compared to the decided difference between normal and abnormal effervescent steel and between normal and abnormal killed steel.

TABLE 4.—*Relative tendencies of normal and abnormal steel, both killed and effervescent, toward the formation of soft spots upon hardening after carburizing*[1]

[The index number has the same significance as in Table 2]

Group No.	Temperature of water	Index number of soft spots			
		Killed steel[2]		Effervescent steel[2]	
		Normal	Abnormal	Normal	Abnormal
	°C.				
1	8	0	93	33	58
2	12	18	58	8	136
3	18	2	43	23	29
4	19	8	61	9	49
5	21	44	103	12	89
6	23	18	88	17	87
7	30	4	97	31	57
8	19	17	112	10	193
9	19	25	128	17	192
10	23	14	106	16	130
11	25	12	185	11	136
12	27	6	192	13	210
13	31	6	126	60	203

[1] The four specimens on each horizontal line in the table were quenched together and constitute a group. The specimens used were carburized blocks 1½ inches square, 2½ inches long. They were heated at 775° C. for 30 minutes in a cyanide salt bath and quenched in water through which oxygen had previously been bubbled for 10 minutes.

[2] The compositions were: Killed steel—C, 0.14 per cent; Mn, 0.44 per cent. Effervescent steel—C, 0.14 per cent; Mn, 0.42 per cent.

(c) SPECIAL MANGANESE CARBURIZING STEEL

Of considerable practical importance were the quenching tests and hardness surveys of special manganese carburizing steels. The measurements of the depth of carburized layer (Table 1) should also be noted in this connection. These quenching tests gave the interesting result that the special manganese carburizing steels are much less prone to give soft spots than even normal plain carbon steel, as may be seen from Tables 5 and 6.

TABLE 5.—*Relative tendencies of special manganese carburizing steel and plain carbon normal steel toward the formation of soft spots upon hardening after carburizing* [1]

[The index number has the same significance as in Table 2]

Pair No.	Temperature of water	Index number of soft spots		Pair No.	Temperature of water	Index number of soft spots	
		Special manganese carburizing steel J [2]	Normal S. A. E. 1020 steel [2]			Special manganese carburizing steel J [2]	Normal S. A. E. 1020 steel [2]
	°C.				°C.		
1	15	0	50	7	19	0	46
2	15	0	8	8	19	0	51
3	17	0	19	9	22	0	18
4	17	0	42	10	22	0	25
5	18	1	30	11	22	2	50
6	18	0	12	12	22	0	78

[1] The size of the specimens used was similar to that described in Table 4; the quenching treatment was also similar.

[2] The compositions were: Special manganese carburizing steel J—C, 0.16 per cent; Mn, 1.36 per cent; P, 0.018 per cent; S, 0.098 per cent; Si, 0.07 per cent. Normal S. A. E. 1020 steel—C, 0.22 per cent; Mn, 0.53 per cent; P, 0.018 per cent; S, 0.098 per cent; Si, 0.07 per cent.

TABLE 6.—*Relative tendencies of special manganese carburizing steels and plain carbon normal and abnormal steels toward the formation of soft spots upon hardening after carburizing* [1]

[The index number has the same significance as in Table 2]

Group number	Temperature of water	Index number of soft spots			
		Special manganese carburizing steel J [2]	Special manganese carburizing steel M [2]	Normal S. A. E. 1020 steel [2]	Abnormal steel
	°C.				
1	24	0	0	17	114
2	27	0	0	15	83
3	29	0	0	29	160
4	31	0	0	25	143
5	32	0	0	4	130

[1] The size of the specimens used was similar to that described in Table 4; the quenching treatment was also similar. Two kinds of special manganese carburizing steel from two different manufacturers were used; one has been designated "J" and the other "M" throughout this paper.

[2] The compositions were: Special manganese carburizing steel J—see Table 5. Special manganese carburizing steel M—C, 0.19; Mn, 1.04; P, 0.060; S, 0.062; Si, 0.13. Normal S. A. E. 1020 steel—see Table 5. Abnormal steel—C, 0.14 per cent; Mn, 0.43 per cent.

As shown in Table 6, under the conditions of quenching used abnormal steel gave a large number of soft spots, normal S. A. E. 1020 steel gave fewer soft spots, but still a considerable number, whereas neither of the special manganese carburizing steels gave any soft spots. Quenching in water through which oxygen had been bubbled is without doubt as favorable toward the formation of soft spots as the quenching procedure in any reasonably careful heat-treating establishment. The conclusion may be drawn, therefore, that in practice complete freedom from soft spots may be obtained by the use of the special manganese carburizing steels. This would also be true, of course, with any other steel containing alloying elements which conferred deep hardening properties equivalent to that of the manganese in the special manganese carburizing steel.

IV. IZOD IMPACT TESTS

As was indicated at the outset, there appears to be no difference in chemical composition, as ordinarily determined, between normal and abnormal steel, nor is it usually considered that there is a difference between them in mechanical properties. However, McKinney[19] has voiced what is probably the general opinion, in stating that if we consider abnormality to be due to the obstruction offered to perfect crystallization and to grain growth, by particles of oxides, we must regard the quality of abnormal steel in respect to mechanical properties as inferior to normal steel, or at least with suspicion, because of the presence of the oxides. On the other hand, because of its finer-grain size, abnormal steel might be presumed to have higher impact strength than normal steel, and this was very strikingly shown to be the fact by a series of impact tests of normal and abnormal carbon steels, and the special manganese carburizing steel. The specimens were tested after various heat treatments. In Table 7 the results given in the last column under "carburizing treatment" are of principal interest. The "carburizing treatment" consisted in heating the specimens for eight hours at 940° C., as for carburizing, no carburizer being used, however. Tests of these specimens therefore, indicated how tough or brittle the cores of these different steels would be after carburizing. The abnormal specimens gave much higher impact resistance after the "carburizing treatment" than the normal specimens. The special manganese carburizing steels also showed high impact resistance after the "carburizing treatment." The above results were confirmed by another series of tests, as shown in the second column of Table 8.

Brittleness of the core induced by the long-continued heating during carburizing may be removed and toughness restored by the "regenerative treatment"; that is, by quenching from the critical range and drawing back. After this treatment, as shown in the last column in Table 8, the impact strength of the normal and abnormal specimens was the same. Nevertheless, from the viewpoint of toughness of the core, the fine-grain size of abnormal steel is an undoubted advantage; for instance, a wider latitude is allowable in the temperature range of heat treatment. It is doubtless of some practical importance that, with abnormal steel, the regenerative treatment may be more safely dispensed with than with normal steel. For these reasons a metallurgist would probably prefer a finer grained steel to the extreme type of normal steel with very large grain. The fine-grained alloy steels commend themselves for the additional reason that, with these, trouble from soft spots is not likely to occur.

[19] P. E. McKinney, "Discussion of normal and abnormal steels," Tr. Am. Soc. Steel Treat., 12, p. 427, 1927.

That the practical metallurgist has not been slow to realize these advantages of fine-grain size, at least in the alloy carburizing steels, is well brought out in the reference already given.[20]

It was of interest to test the impact resistance of the core of normal and abnormal steel after the quenching for hardening the case. Bars 1½ inches square cross section were given the "carburizing treatment" and quenched from 775° C. into water, as for casehardening. Standard Izod impact specimens, 1 cm by 1 cm in cross section, were then cut from the bars one-eighth inch and one-half inch below the surfaces. The results are shown in Table 9. The much higher impact resistance of the abnormal steel after this treatment was striking. It was an object of the tests to determine whether or not the high manganese content in the special manganese carburizing steels tended to make the core brittle, after the quenching to harden the case. Although the impact resistance of the two special manganese carburizing steels in these tests was much lower than that of the abnormal steel, it was not apparent that the cores of the special manganese carburizing steels were more brittle than the core of the normal steel. It should be noted that no tests were made in this series in which the regenerative treatment was given specimens after the "carburizing treatment" and previous to the quenching.

TABLE 7.—*Impact resistance* [1] *of normal and abnormal steel and special manganese carburizing steel after various heat treatments*

Specimen No.[2]	Heat treatment				
	As received		Normalized—heated 1 hour at 900° C. and air cooled		Carburizing treatment [3]
	Energy absorbed, in foot-pounds				
Normal steel:					
122	74	73	60	58	23
134	78	59	76	67	23
146	77	77	77	79	31
152	80	82	77	71	42
155	23	23	72	71	17
Abnormal steel:					
113	74	76	77	79	74
119	74	77	77	78	69
125	73	72	75	75	69
137	76	77	78	78	74
143	79	81	80	80	59
149	83	79	82	85	69
Special manganese carburizing steel J	62	63	78	77	70

[1] A 120-foot-pound capacity Izod machine was used. The specimens were triple notched, and each value for energy absorbed is the average for the 3 notches.
[2] For the compositions of these specimens, see Table 1.
[3] By "carburizing treatment" is meant heating for 8 hours at 940° C., and cooling in the furnace as in carburizing (no carburizer being used, however), after which the specimens were machined so that, if any oxidation occurred during the heating, the oxidized surfaces were removed.

[20] See footnote 18, p. 439. See also discussion by H. W. McQuaid, Trans. Am. Soc. for Steel Treat., 13, p. 1002; 1928.

TABLE 8.—*Impact resistance* [1] *of normal and abnormal steel and special manganese carburizing steel after various heat treatments*

Specimen No.[2]	Heat treatment					
	As received	Carburizing treatment [3]		Carburizing followed by treatment [4]		Treatment regenerative [3]
	Energy absorbed, in foot-pounds					
Normal steel:						
1M	72	16	21	85	79	77
8M	56	20	25	84	79	76
9M	72	24	26	89	81	84
11M	66	20	23	86	81	45
12M	73	30	36	92	84	87
14M	57	21	26	86	77	74
15M	73	19	31	90	81	83
18B	74	34	36	88	83	85
19M	75	20	19	93	86	89
21M	74	25	24	93	85	89
43	74	32	38	101	92	94
122	75	40	39	99	94	97
134	75	22	27	96	88	91
146	78	25	34	96	88	88
152	79	42	51	103	102	98
155	4	19	24	93	85	89
Abnormal steel:						
10T	70	36	40	86	59	74
13B	76	66	62	91	79	83
16M	73	45	46	87	80	84
113	79	73	61	97	91	96
119	71	62	60	110	96	104
125	72	69	74	94	89	96
131	73	35	55	102	95	92
137	76	67	71	95	88	91
143	81	54	76	101	95	98
149	82	54	62	115	109	114
E	68	32	45	89	75	----------
Special manganese carburizing steel J	51	61	70	81	78	78

[1] A 120-foot-pound capacity Izod machine was used. The specimens were triple notched, and each value for energy absorbed is the average for the three notches.

[2] The chemical compostion of some of these specimens is given in Table 1, the compostion of the rest is given below:

Specimen No.	C	Mn	P	S	Si	Ni
	Per cent	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*
1M	0.12	0.37	0.012	0.020	0.01	----------
8M	.11	.31	.013	.023	----------	----------
9M	.14	.31	----------	----------	.09	----------
11M	.12	.38	.012	.025	.01	----------
12M	.13	.41	----------	----------	.10	----------
14M	.12	.34	.010	.027	.005	----------
15M	.13	.36	----------	----------	.07	----------
18B	.12	.39	.013	.038	.08	----------
19M	.14	.35	.010	.024	.074	----------
21M	.12	.40	.012	.020	.008	----------
43	.17	.47	.018	.027	.14	0.18
10T	.12	.34	.013	.023	.01	----------
13B	.12	.39	.012	.025	.004	----------
16M	.13	.37	.010	.027	.008	----------

[3] By "carburizing treatment" is meant heating for eight hours at 940° C., and cooling in the furnace as in carburizing (no carburizer being used, however), after which the specimens were machined so that, if any oxidization occurred during the heating, the oxidized surfaces were removed.

[4] Regenerative treatment—oil quenching from 870° C., and drawing at 700° C.

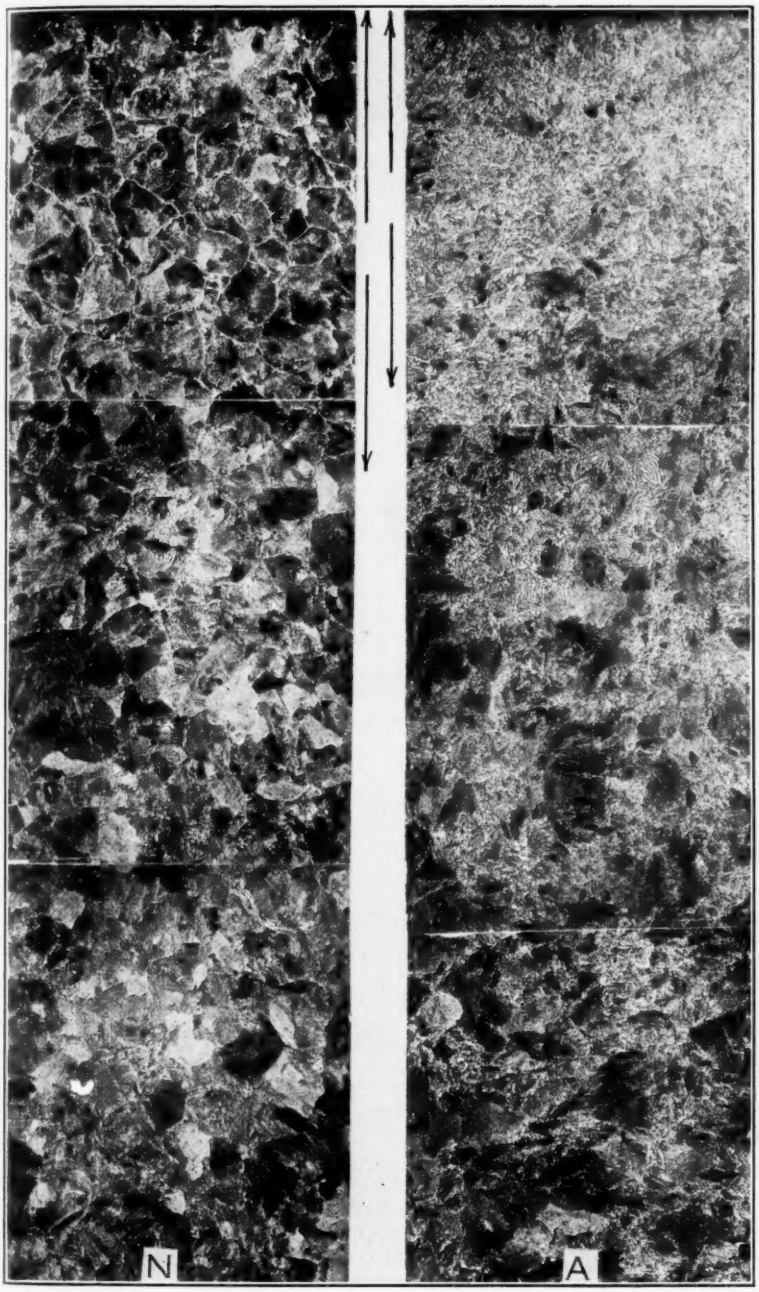

FIG. 13.—*Structure of the carburized layers of normal and abnormal tool-steel specimens.* × *100*

The same characteristics of the normal and abnormal structure appear in the tool steel as in the carburizing steel. The arrows indicate, approximately, the extent of the hypereutectoid zone

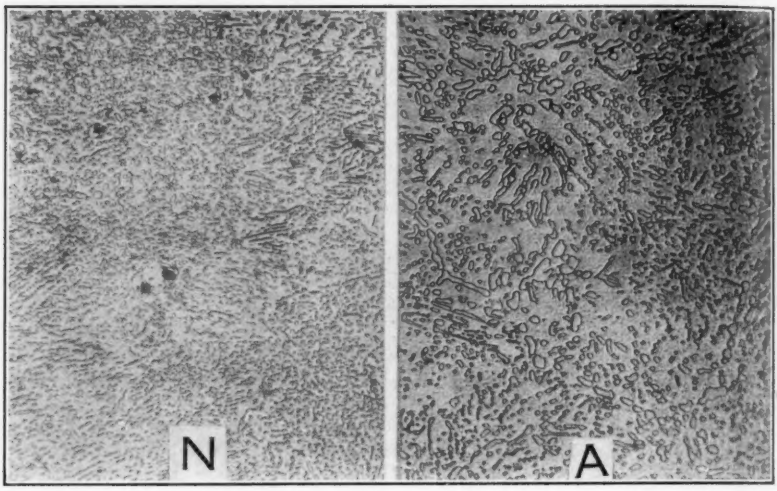

FIG. 14.—*The structure of normal and abnormal tool-steel specimens after cooling slowly through the critical range.* × *500*

The coalescence or spheroidization of the cementite has progressed further in the abnormal steel than in the normal steel. The specimens were cooled together

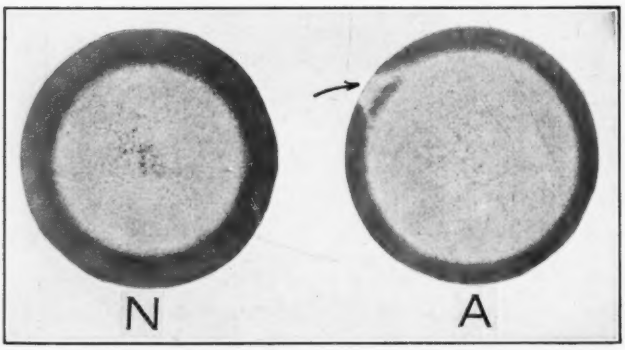

FIG. 15.—*Cross sections of quenched normal and abnormal tool-steel cylinders; deeply etched with hot 1 : 1 hydrochloric acid.* × *1*

The dark outer ring shows the depth of hardening. The arrow indicates a soft spot in the abnormal specimen. The two specimens were quenched simultaneously in the same bath

TABLE 9.—*Impact resistance* [1] *of normal and abnormal steel and special manganese carburizing steel, after quenching* [2]

Specimen No.[3]	⅛ inch from surface		½ inch from surface	
	Energy absorbed, in foot-pounds			
Normal steel:				
122	9	11	30	13
159	31	14	39	29
Abnormal steel: 113	76	62	88	77
Special manganese carburizing steel M	9	10	25	25
Special manganese carburizing steel J	22	14	38	24

[1] A 120-foot-pound capacity Izod machine was used. The specimens were triple notched, and each value for energy absorbed is the average for the three notches.

[2] Bars 1½ inches square, 6 inches long, were water quenched, from 775° C. after being given the "carburizing treatment." Izod impact specimens were then cut longitudinally from the bars, 1 impact specimen ⅛ inch from the surface and another ½ inch from the surface.

[3] The chemical compositions of these steels is given below:

Specimen No.	C	Mn	P	S	Si
	Per cent	*Per cent*	*Per cent*	*Per cent*	*Per cent*
122	0.14	0.44			
159	.22	.53	0.018	0.048	0.02
113	.14	.42			

Special manganese carburizing steel J } See Tables 5 and 6.
Special manganese carburizing steel M }

V. NORMAL AND ABNORMAL TOOL STEEL

The characteristics of the normal and abnormal structure present in low-carbon carburizing steel have also been found to exist in high-carbon tool steel.[21] In order to bring out the differences between the normal and abnormal types, it is necessary to make the McQuaid-Ehn carburizing test with the tool steels just as with the carburizing steels.[22] Figure 13 shows the structures of a normal and abnormal tool steel of the following composition: Normal—C, 0.90 per cent; Mn, 0.28 per cent; Cr, 0.15 per cent. Abnormal—C, 0.90 per cent; Mn, 0.23 per cent; Cr, 0.01 per cent. The presence of the 0.15 per cent chromium in the normal steel (which apparently did not represent an intentional addition of chromium, the steel having been submitted as a plain carbon tool steel) makes it somewhat uncertain whether the more uniform hardening of the normal steel shown in Table 10 was due entirely to its structural normality or partly to the chromium content. In the above abnormal steel the tendency for coalescence of the cementite in the hypereutectoid zone manifested itself, after suitable treatment, also in the interior of the steel. Figure 14 shows the structure of samples of these normal and abnormal tool steels cooled together slowly through the critical range. It can

[21] F. G. Sefing, "Abnormal *v.* Normal Tool Steels," Michigan State College, Eng. Exper. Station, Bull. No. 5.

[22] Acknowledgment is due to the Chicago Pneumatic Tool Co. and the Carpenter Steel Co. for furnishing specimens of tool steel.

be seen that in the abnormal steel the coalescence or spheroidization of the cementite has progressed further than in the normal steel. In this connection the thought suggests itself that the variation in the response to spheroidizing treatments of different samples of carbon steel of similar composition, noted by Whitley,[23] may have been connected with the normality or abnormality of the specimens used

In quenching tests of the above tool steels the abnormal samples showed themselves more prone to give soft spots than the normal samples; 1½-inch blocks similar to the carburized blocks could not be obtained, and instead smaller samples, 1¼-inch rounds, 2 inches long, were used. These gave fewer soft spots than the larger carburized blocks, but, as shown in Table 10, the indications were that the abnormal samples were more prone to give soft spots. Brine quenching was not used with the tool steels, since it was evident from the preceding work that complete hardening would be obtained in this way.

TABLE 10.—*Index numbers*[1] *of soft spots of normal and abnormal tool steels*[2]

Pair No.	Heating bath	Quenched		Index number of soft spots	
		In—	At—	Normal	Abnormal
			° C.		
1	Air atmosphere	Tap water	29	0	31
2	do	do	28	1	30
3	do	do	17	0	3
4	do	do	17	0	0
5	do	do	17	1	1
6	Cyanide salt bath	do	28	0	15
7	do	do	28	0	14
8	do	do	20	0	0
9	do	do	19	0	2
10	do	do	28	0	22
11	do	do	28	0	3
12	do	Tap water plus oxygen	28	0	7
13	do	do	28	0	24
14	do	do	30	0	20
15	do	do	30	0	30

[1] 1¼-inch rounds, 2 inches long, were used. Ninety-six Rockwell readings were taken on each specimen, as with the carburized blocks. The specimens were heated for 30 minutes at 775° C. before quenching.
[2] The composition is given below:
Normal: C, 0.90 per cent; Mn, 0.28 per cent; Cr, 0.15 per cent.
Abnormal: C, 0.90 per cent; Mn, 0.23 per cent; Cr, 0.01 per cent.

Figure 15 shows cross sections of quenched samples of the above normal and abnormal tool steels deeply etched with 1:1 hydrochloric acid. The dark outer ring in each specimen shows the depth of the hardened layer. The normal steel hardened considerably more deeply than the abnormal steel. An arrow indicates a soft spot in the abnormal steel. Here again it is uncertain whether the greater depth of hardening of the normal steel was due entirely to its struc-

[23] J. H. Whitely, "The formation of globular pearlite," Iron & Steel Inst., **105**, p. 346, 1922.

tural normality or partly to its chromium content. It is to be regretted that specimens of normal and abnormal tool steel of exactly similar composition could not be obtained for these tests. At any rate it appears that, on account of the possible variations in the content of carbon, manganese, chromium, etc., in carbon tool steels, the chemical composition may be as important in determining the critical cooling rate necessary for complete hardening as the structural normality.

Practically, as has already been pointed out for carburized steels, different degrees of normality may be desired also for tool steels, depending on the purpose for which they are to be used. For example, for penumatic tools a less deeply hardening steel is preferred, to counteract brittleness.

The experimental work so far described shows that there exist in steels differencies, of which the most sensitive indication is the microstructure of the carburized layer. A definition of abnormality could, however, be made on the basic of the critical cooling rate necessary for hardening after carburizing. The connection between abnormality thus defined and soft spots would then be obvious—no soft spots would occur if the critical cooling rate were exceeded. The microstructure and the critical cooling rate are undoubtedly two different manifestations of the same underlying cause.

VI. CAUSES OF ABNORMALITY

According to Ehn, abnormality is caused by dissolved or submicroscopic particles of oxides. In discussion of his work other causes were suggested, such as high nitrogen content and high phosphorus content. Indeed, in tests at the Bureau of Standards it has been found that almost any pronounced irregularity in chemical composition of the specimen is plainly reflected in the appearance of the carburized structure. Thus, the structure of wrought iron, of electrolytic iron, of highly phosphorized, oxidized, or nitrogenized iron, of the fused-in metal of welds,[24] of small steel ingots low in manganese melted in a laboratory induction furnace, and of unfinished steel sampled in the progress of a heat, all showed marked differences after carburizing from the structures of normal steel. From this viewpoint there may be any number of causes for structural abnormality of steel. It must be noted, however, that most of the above-mentioned causes would hardly apply to commercial steel.

In commercial steel we need consider as possible causes of abnormality only such factors as are present in everyday steel making; we must be careful not to infer too readily that because a thing causes abnormality in a laboratory sample it is consequently a cause of abnormality in commercial steel. For instance, in laboratory melts

[24] See footnote 3, p. 425.

excessive phosphorus content causes an abnormal structure. It does not follow, however, that phosphorus is a cause of abnormality in commercial steel; actually, as shown by chemical analyses of commercial normal and abnormal steel, there seems to be no relation between the phosphorus content and abnormality. Again, in laboratory melts low manganese content, especially when accompanied by low carbon content,[25] gave an abnormal structure. Low manganese content, however, can hardly be regarded as a cause of abnormality in commercial steel, for, as shown by chemical analyses of ordinary commercial normal and abnormal steels, there seems to be no obvious relation between the manganese content and abnormality. Similarly, it was found in laboratory experiments[26] that excessive oxygen content caused an abnormal structure. Here, also, we must realize we are not justified in concluding that abnormality is caused by high oxygen content, unless analyses of representative abnormal steel specimens actually show a high oxygen content. It is not enough to show that some abnormal steel has a high oxygen content, for such specimens may be exceptional. To justify the conclusion that abnormal steel is caused by high oxygen content, it must be shown that the general run of abnormal steel is high in oxygen. Due regard for these considerations would help remove much of the existing confusion as to the causes of abnormality in commercial steel.

The chief difficulty involved in the subject of the relation between abnormality and oxygen content is due to the imperfect state of the methods of oxygen analysis. Ehn did not try to show any quantitative relation between abnormality and oxygen content but suggested instead that abnormality is not due primarily to the amount of oxides present in the steel, but to their type and form—in particular, that the oxides effective in causing abnormality are present in solution or as colloidal or submicroscopic particles. In this form, Ehn's explanation of abnormality is not readily capable of quantitative verification. In spite of the decided improvements that have recently been made in the methods of oxygen analysis, it has not yet been possible to find a quantitative relation between abnormality and oxygen content. In the present work, during the gathering of data bearing on the causes of abnormality, nothing has arisen to weaken Ehn's theory; on the contrary, the data tend rather to support it, and, although not definitely established, it continues to be a helpful guiding principle. Below are given the results of the work done in this investigation to determine the cause of abnormality.

[25] Unpublished results by the Bureau of Mines.

[26] In tests at the Bureau of Standards electrolytic iron, remelted in an induction furnace in air, showed an extremely abnormal structure. See also Harder, Weber, and Jerabek, footnote 9, p. 430; they found that steel heated in oxygen became very abnormal.

1. SPECTROGRAPHIC ANALYSES FOR ALUMINUM

Ordinary chemical analyses have failed to show any relation between abnormality and the amounts of the usual elements determined in steel—namely, sulphur, manganese, phosphorus, and silicon—so that abnormality can hardly be ascribed to variations in content of these usual elements. A relation was found, however, between aluminum content (probably more strictly, alumina content) and abnormality. It had been observed in a rather extensive series of preliminary experiments [27] that steel treated with aluminum in the mold was abnormal. In spectrographic analyses [28] of a large number of commercial normal and abnormal steel specimens it was likewise found that abnormality was generally associated with high aluminum content. Figure 16 shows the results of these spectrographic analyses. The intensity of the aluminum lines in the arc spectrum of the specimen is proportional to the aluminum content. With the aid of chemical analyses [29] of a series of specimens of varying aluminum content, a given intensity of the aluminum lines in the spectrum was

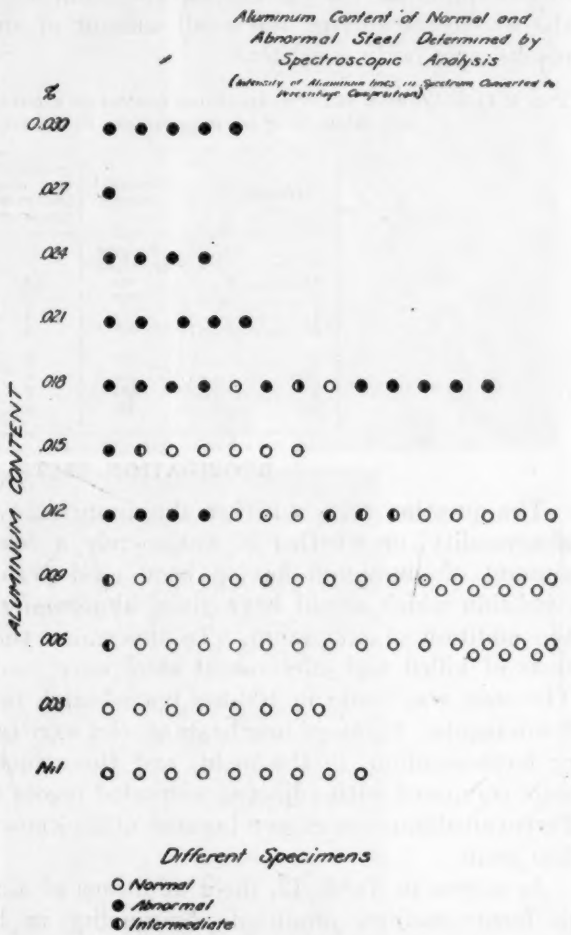

FIG. 16.—*Results of spectrographic analyses for aluminum of normal and abnormal steel specimens*

About half of the specimens were commercial steels; the other half, though experimental, were made on a commercial scale in 100-ton furnaces.

[27] Acknowledgment is due to the Trumbull Steel Co., at whose plant the experiments were made, and to the Titanium Alloy Manufacturing Co. and the Union Carbide & Carbon Co., who cooperated.

[28] Spectrographic analyses by W. F. Meggers, physicist, Bureau of Standards.

[29] H. A. Bright, associate chemist, Bureau of Standards.

converted to percentage composition of aluminum. Unfortunately for the general utility of the spectrographic method, it appears that consistent results can be obtained only after much experience.

Table 11 shows the interrelation between the results of the chemical analyses for aluminum of a number of specimens and the intensity of the aluminum lines in the arc spectrum of these specimens. On the whole, considering the small amount of aluminum present, the results were fairly consistent.

TABLE 11.—*Relation between aluminum content as determined by chemical analysis and intensity of aluminum lines in the arc spectrum*

Specimen No.	Chemical analysis	Intensity of aluminum lines in arc spectrum
	Per cent aluminum	
54	0.009	3
2	.010	3
1	.010	3
15	.011	4
47	.016	6
10B	.026	6
E	.020	8
22	.027	8

2. DEOXIDATION TESTS

The question arose whether aluminum was actually the cause of abnormality, or whether it was merely a concomitant, the larger amount of aluminum having been used perhaps to counteract a condition which would have given abnormal steel with or without the addition of aluminum. To determine this point, deoxidation tests of killed and effervescent steel were made at a steel plant.[30] The steel was made in 100-ton open-hearth furnaces and cast into 3-ton ingots. Ingots of four heats of steel were treated with aluminum or ferrovanadium in the mold, and the structures of these ingots were compared with adjacent untreated ingots from the same heats. Ferrovanadium was chosen because of its known effect in producing fine grain.

As shown in Table 12, mold additions of aluminum and likewise of ferrovanadium produced abnormality in both the killed and effervescent steel. The additions of aluminum and ferrovanadium used in the mold were just sufficient to kill the ingots of effervescent steel. Figures 17, 18, 19, and 20 show micrographs of the carburized layers of bars from the regular ingots of the killed and effervescent steels and of bars from ingots of the same heats, treated with aluminum or ferrovanadium in the mold. The samples from the untreated ingots were normal; those from the treated ingots were abnormal. In the carburized layers of the untreated effervescent steel samples

[30] These tests were made possible through the courtesy of the Central Alloy Steel Corporation, Canton Ohio.

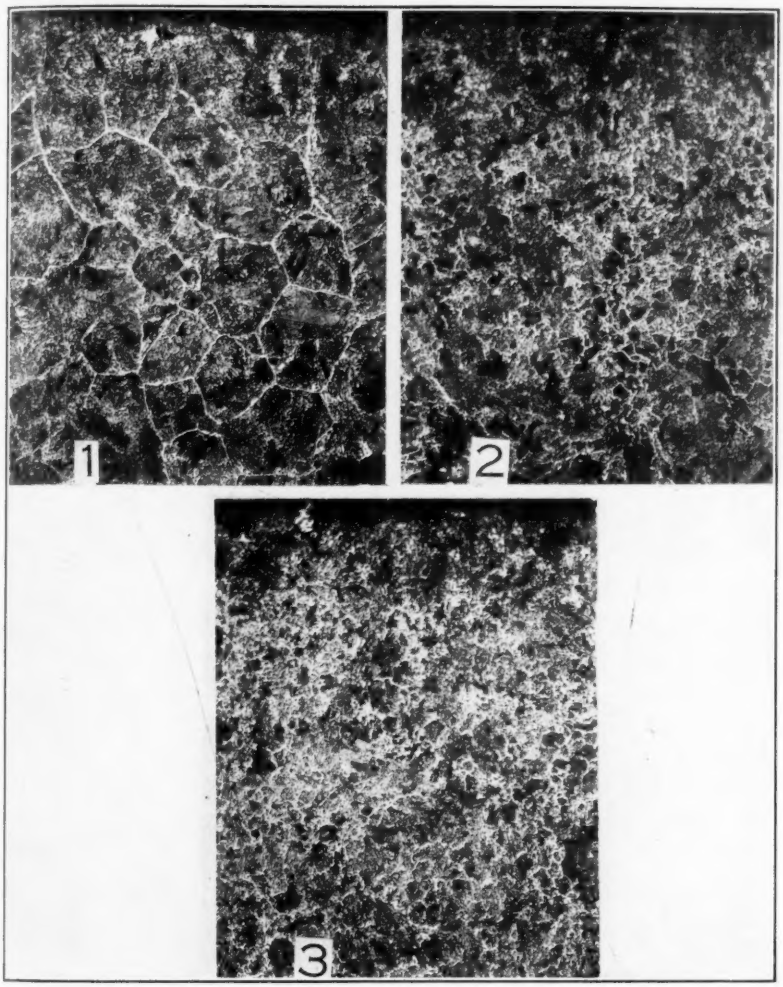

FIG. 17.—*Structure of low carbon (0.11 to 0.14 per cent) "killed" steel after carburizing.* × *100*

The steel was killed by the use of ferrosilicon; *1*, forged down from regular commercial ingot; normal; *2*, forged down from 3-ton ingot from same heat with 4 pounds of shot aluminum added in mold; abnormal; *3*, forged down from 3-ton ingot from same heat with 25 pounds of 37 per cent ferrovanadium added in mold; abnormal

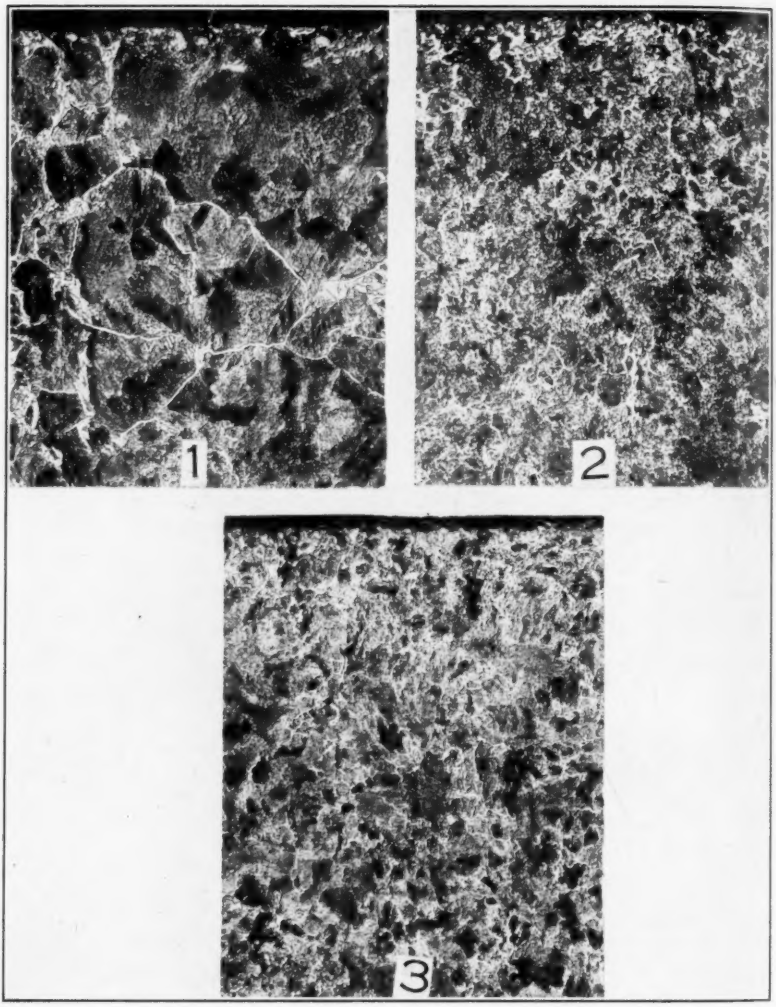

FIG. 18.—*Structure of low carbon (0.11 to 0.14 per cent) effervescent steel after carburizing.* × *100*

1, forged down from regular commercial ingot; normal; *2*, forged down from 3-ton ingot from same heat with 4 pounds of shot aluminum added in mold; abnormal; *3*, forged down from 3-ton ingot from same heat with 25 pounds of 37 per cent ferrovanadium added in mold; abnormal

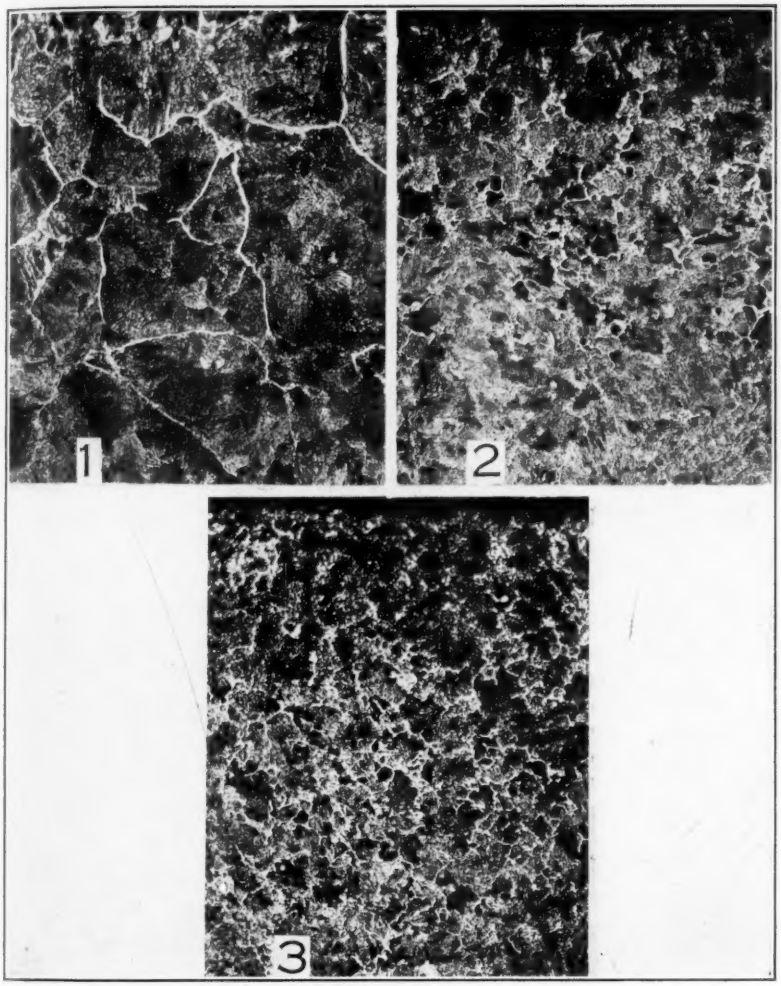

FIG. 19.—*Structure of low carbon (0.11 to 0.14 per cent) effervescent steel after carburizing.* × *100*

1, forged down from regular commercial ingot; normal; *2*, forged down from 3-ton ingot from same heat with 5 pounds shot aluminum added in mold; abnormal; *3*, forged down from 3-ton ingot from same heat with 25 pounds of 37 per cent ferrovanadium and 1½ pounds of shot aluminum added in mold; abnormal

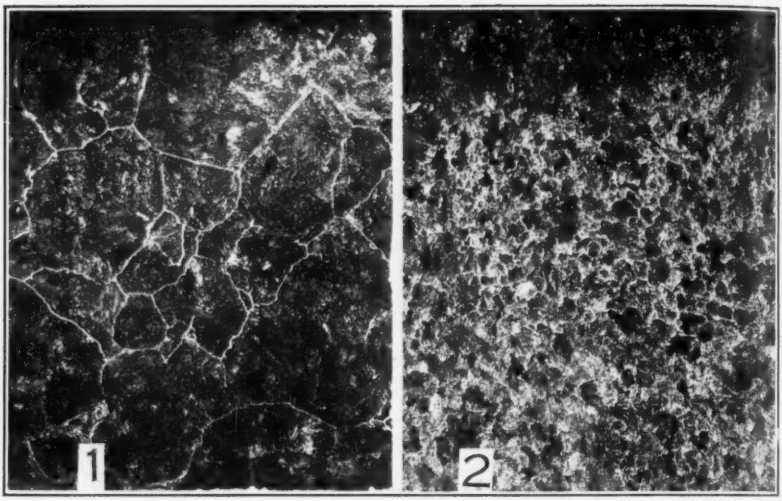

FIG. 20.—*Structure of one-half per cent nickel "killed" steel after carburizing.* × *100*

1, forged down from regular commercial ingot; normal; *2*, forged down from 3-ton ingot from same heat with 25 pounds of 37 per cent ferrovanadium added in mold; abnormal

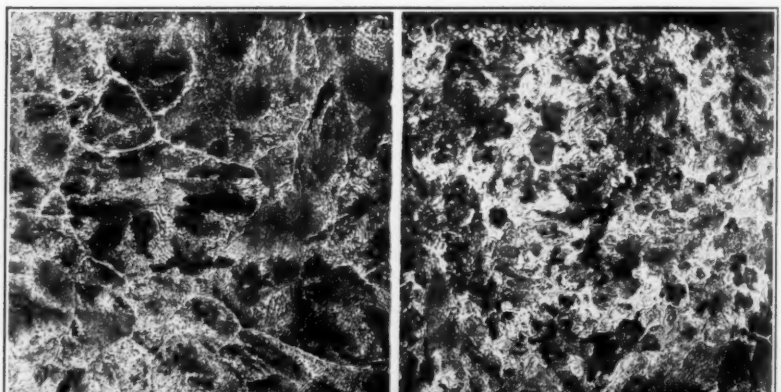

FIG. 21.—*Structure of the carburized layers of two ingots of 0.12 per cent carbon effervescent steel from the same heat, the ingots deoxidized with silicon in the mold and aluminum in the mold, respectively.* × *100*

The silicon treated steel is normal; the aluminum treated steel is abnormal

the cementite envelopes were somewhat thicker than in the untreated killed steel samples; the grain size was large, however, with no marked separation of the cementite from the ferrite, so that it appears justifiable to classify the untreated effervescent samples as normal. The important point was noted that in the killed steel about the same amount of aluminum per ton of steel was added in the ladle, as a matter of regular practice, as was used in the experimental mold additions to the killed and effervescent steels. However, in contrast to the mold additions, the ladle additions apparently did not produce abnormality. The effect of either the aluminum or ferrovanadium additions appeared to be the same with the killed steels, to which aluminum had already been added in the ladle, as with the effervescent steels, to which no aluminum additions had previously been made.

TABLE 12.—*Effect on degree of normality of additions of aluminum or ferrovanadium in the mold to successive ingots of heats of killed and effervescent steel*

Type of steel	Ingot No.	Mold addition to 3-ton ingot	Structure of carburized layer
Heat 1, 0.11 to 0.14 per cent C, killed steel	7	4 pounds shot Al	Abnormal.
	8	do	Do.
	9	25 pounds 37 per cent ferrovanadium	Do.
	10	Regular ingot, no mold addition	Normal.
Heat 2, 0.11 to 0.14 per cent C, effervescent steel.	7	4 pounds shot Al	Abnormal.
	8	do	Do.
	9	25 pounds 37 per cent ferrovanadium	Do.
	10	Regular ingot, no mold addition	Normal.
Heat 3, 0.11 to 0.14 per cent C, effervescent steel.	7	5 pounds shot Al	Abnormal.
	8	do	Do.
	9	25 pounds 37 per cent ferrovanadium 1½ pounds shot Al.	Do.
	10	Regular ingot, no mold addition	Normal.
Heat 4, 0.13 to 0.16 C, 1½ per cent Ni, killed steel.	7	25 pounds ferrovanadium	Abnormal.
	8	Regular ingot, no mold addition	Normal.

In previous tests it was found that the effect of silicon was quite different from that of aluminum. Figure 21 shows micrographs of the carburized layers of two ingots of 0.12 per cent carbon effervescent steel from the same heat, the ingots killed in the mold with ferrosilicon and aluminum respectively. The silicon-treated steel was normal; the aluminum-treated steel was abnormal. The idea should not be gained, however, that a very low silicon content, say below 0.01 per cent, such as is found in ingot iron and in much effervescent steel, in itself produces abnormality, and that the presence of larger amounts of silicon produces normality. Perusal of the tables of chemical compositions given in this paper will show that this is not true; for example, in Table 13 steels A and E, with over 0.20 per cent silicon, were abnormal and steel R, with only 0.002 per cent silicon, was normal.[31]

[31] Steels (High Silicon Structural Steel, H. W. Gillett, B. S. Tech. Paper No. 331, p. 129, 1926) of about 0.11 per cent C, 0.90 per cent Si, with about 0.50 to 0.80 per cent Mn, containing as well 0.02 to 0.03 per cent Al., were classed as intermediate or fine-grained normal.

The results given in Table 12 clearly indicate that abnormality, if it is not caused directly by oxides in the steel, is at least closely associated with the manner of deoxidation. It is significant that ladle additions of aluminum did not produce abnormality, whereas mold additions did. In both instances a large proportion of the aluminum may be presumed to have become oxidized, and it appears very probable that the difference in effect may be due to a difference in the form or size of the oxide particle, the finer particles causing abnormality. When aluminum is added in the large 100-ton ladle the aluminum oxide has a chance to coalesce into larger particles and rise to the slag. When aluminum is added in the mold, however, the oxide particles are more likely to remain in a fine state of subdivision than when the addition is made earlier; that is, in the ladle. The difference in action between silicon and aluminum may, perhaps, be accounted for similarly. The oxides formed from the addition of silicon were probably more fusible than those formed from the addition of aluminum and may, therefore, more readily coalesce into larger particles. It has also been suggested that perhaps the reason why the addition of aluminum in the ladel did not produce abnormality was that this aluminum, instead of forming alumina, reacted to form aluminum silicates with the regular addition of ferrosilicon which was added in the ladle at the same time.

An interesting observation of the deoxidation tests was that additions of aluminum gave similar results in respect to abnormality as additions of ferrovanadium. The effect of vanadium in producing fine grain is well known and is probably largely due to obstruction to grain gowth by fine particles of vanadium compounds, perhaps oxides or carbides. The vanadium-treated steels gave the same number of soft spots in quenching tests and hardness surveys as the aluminum-treated steels. Aluminum-treated abnormal specimens showed the same superior impact resistance to normal steel, as ferrovanadium treated abnormal steel. (See Sec. IV.) The suggestion presents itself that the effect of an expensive alloying element, such as vanadium, in producing the desirable property of fine grain, may be simulated by inexpensive additions of aluminum to both alloy and plain carbon steel. Of course, the disadvantage of aluminum is the tendency of its oxides to segregate, but this might, perhaps, be counteracted by adding the aluminum in less concentrated form, perhaps as an alloy with iron, and in the minimum amount that will produce the effect. However, the uncertainty as to the oxygen content of the steel at the moment of adding the aluminum would no doubt make it very difficult to get dependable results.

3. EXAMINATION OF SMALL TEST INGOTS TAKEN DURING PROGRESS OF OPEN-HEARTH HEATS

The deoxidation tests described above plainly indicate that abnormality may be due to the manner of deoxidation. A series of McQuaid-Ehn tests of very small test ingots taken during the progress of open-hearth heats appeared to indicate that in finished steel the deoxidation treatment is the main factor, apparently overshadowing any differences in the degree of normality that may have been produced during the melting and refining of the steel. Figure 22 shows the carburized layers of a series of small test ingots from an open-hearth heat of 0.20 per cent carbon killed steel. It can be seen that samples taken during the refining of the heat showed decided coalescence of the cementite. After deoxidation, however, the steel seemed to be perfectly normal, the degree of normality apparently reflecting mainly the deoxidizing treatment used. Figure 23 shows the carburized layers of a series of small test ingots from another open-hearth heat in which samples taken during the progress of the heat showed very little coalescence of the cementite. This steel, after deoxidation, was likewise normal.

The fact (discussed under II, 5) that variations in the mechanical treatment and heat treatment of steel have practically no effect on the degree of normality indicated that abnormality has its origin in the making of the steel. The experiments above indicated that in the heats of steel tested the degree of normality of the finished steel was more largely dependent on the deoxidation procedure than on the melting procedure. The results obtained in the deoxidation tests with aluminum bear out the results of the spectrographic analyses, which showed that abnormality in commercial plain carbon carburizing steel was almost invariably accompanied by a high aluminum content. From the above, and in view of the common use of aluminum in the deoxidation of steel, the conclusion appears justifiable that most cases of abnormality in commercial steel are due to aluminum oxide

4. OXYGEN ANALYSES

Although abnormality was shown to be connected with the deoxidation procedure, no direct relation was found between oxygen content and abnormality. Table 13 gives the results of gas analyses by the vacuum-fusion method [32] of representative specimens of commercial normal and abnormal steel. There is no noticeable relation between the nitrogen or hydrogen contents and abnormality. To visualize better the results of the oxygen analyses, the specimens in Table 13 have been divided into two groups of lower and higher

[32] L. Jordan and J. R. Eckman, B. S. Sci. Paper No. 514. Gases in Metals, II. The Determination of Oxygen and Hydrogen in Metals by Fusion in Vacuum.

apparent oxygen content, respectively. It can be seen that abnormal specimens do not as a rule show higher apparent oxygen content; on the contrary, most of them fall into Group I of lower apparent oxygen content. The values given are for apparent oxygen content as determined by the original vacuum-fusion method. They give no information about the form in which the oxygen determined is present in the steel, whether in solution or as inclusions; and if as inclusions, whether these are compounds of iron, manganese, or silicon, for example.

TABLE 13.—*Results of gas analyses of normal and abnormal steel specimens* [1]

Specimen No.	Structure of case	Chemical composition					Gas content (uncarburized stock)		
		C	Mn	P	S	Si	N_2 [2]	H_2 [3]	Apparent O_2 [3]
	Group I (lower oxygen content):	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*	*Per cent*
A	Abnormal	0.23	0.52	0.027	0.030	0.23	0.012	(4)	(4)
B	Intermediate	.21	.52	.032	.030	.19	.009	(4)	(4)
C	Abnormal	.46	.71					0.0002	(4)
D	Do	.19	.48	.028	.025	.06	.015	.0001	0.002
E	Do	.21	.51	.024	.029	.22	.011	.0001	.002
F	Intermediate	.21	.43	.009	.042	.006	.005	.004	.003
G	Abnormal	.16	.42				.003		.003
H	Normal	.17	.43					(4)	.002
I	Do	.14	.51					(4)	.003
J	Do	.21	.17					.0001	.003
K	Abnormal	.19	.42	.009	.043	.006	.003	(4)	.003
L	Do	.22	.41						.004
M	Do	.19	.48	.004	.027	.03	.005	.0005	.005
N	Do	.26	.52	.019	.036	.009	.004	.0004	.006
O	Do	.12	.34			.01		.0006	.003
P	Do	.12	.39			.004		.0009	.003
Q	Do	.13	.37			.008		(4)	(4)
	Group II (higher oxygen content):								
R	Normal	.19	.41	.004	.050	.002		.001	.009
S	Do	.16	.44	.009	.019	.05	.005	.0007	.011
T	Do	.25	.46	.014	.027	.13	.004	.0007	.012
U	Do	.17	.60	.012	.021	.03	.004	.0006	.012
V	Abnormal	.08	.33					.0005	.017
W	Do	.07	.37	.014	.021	.013	.010	(4)	.032
X	Normal	.12	.37			.01		.0004	.008
Y	Do	.12	.27			.07		(4)	.037
Z	Do	.12	.28					.0007	.012
AA	Do	.12	.38			.08		.0001	.018
BB	Do	.14	.35			.07		.0001	.017
CC	Do	.14	.25			.07		.0001	.021

[1] Analyses by W. P. Barrows, assistant chemist, and R. J. Kranauer, assistant scientific aid, Bureau of Standards.
[2] Combined nitrogen by the Allen method.
[3] Apparent oxygen and hydrogen by the vacuum-fusion method.
[4] Not detected.

The fact that most of the abnormal specimens in Table 13 showed lower apparent oxygen content than the normal specimens requires further mention. As we have seen, the commercial abnormal steels were generally high in aluminum. It is quite certain that aluminum oxide, which, presumably, was present in these abnormal steels, is

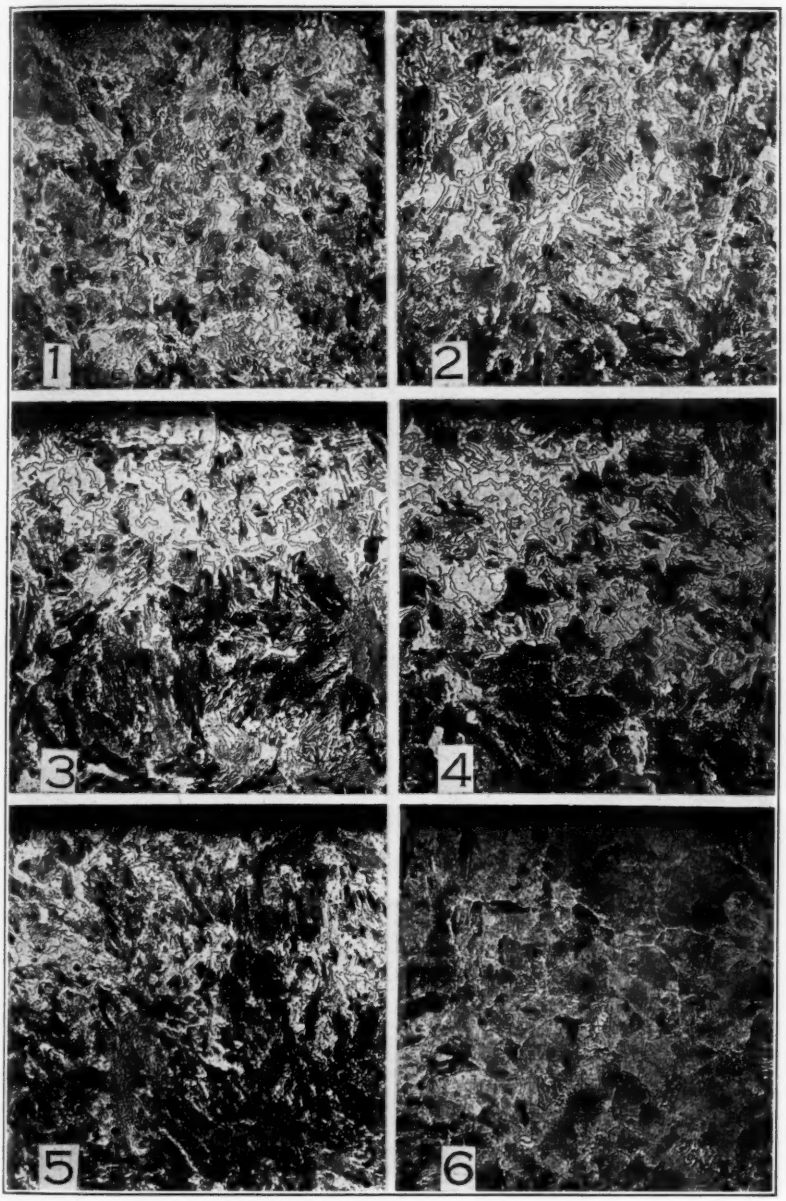

FIG. 22.—*Change in degree of normality in commercial steel accompanying deoxidation.* × *100*

1 to *5*, carburized layers of small test ingots taken during the progress of a 0.20 per cent carbon steel open-hearth heat. These test ingots were taken at half-hour intervals during three hours prior to tapping the heat. The micrographs show that this heat before deoxidation was pronouncedly abnormal; *6*, carburized layer of the finished steel after deoxidation; the structure is normal

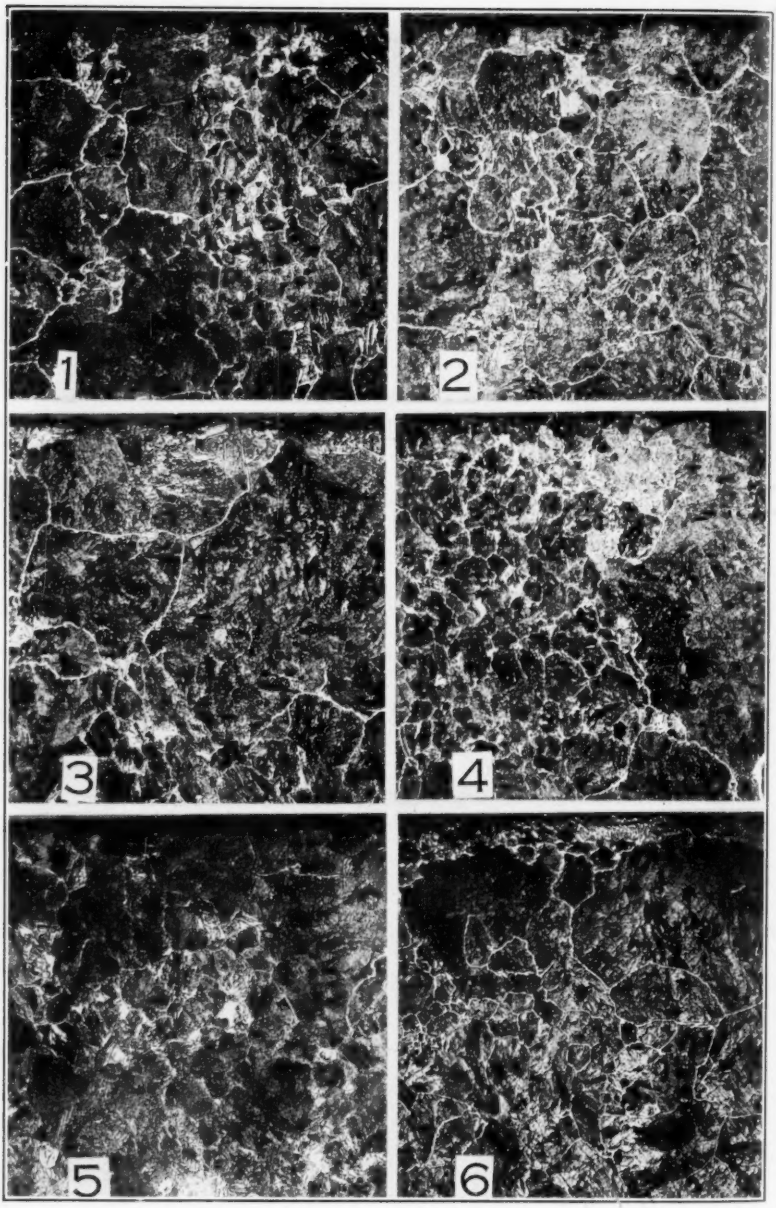

FIG. 23.—*Commercial steel which showed no marked change in normality during deoxidation.* × *100*

1 to *5*, carburized layers of small test ingots taken during the progress of a 0.30 per cent carbon, 0.60 per cent chromium, and 1 per cent nickel steel open-hearth heat. These test ingots were taken at half-hour intervals during three hours prior to tapping the heat. This heat did not appear to go through a pronouncedly abnormal stage, as did the heat shown in Figure 24. The structures *1* to *5* may be called almost normal; *6*, the carburized layer of the finished steel after deoxidation; the structure is likewise normal

not completely reduced during the analysis by the vacuum-fusion method, and it follows that the abnormal specimens of Table 13 probably have a higher total oxygen content than that indicated by the analysis. This is well borne out by oxygen analyses by the vacuum-fusion method of the steels treated with aluminum and ferrovanadium in the mold and of adjacent untreated ingots, as shown in Table 14. After the addition of aluminum in the mold, the apparent oxygen content in the steel was negligible, although the oxygen content of the regular untreated ingots from the same heats was not unusual. It should be noted, in the first two heats, that there was no change in the oxygen content from that of the regular untreated steel when ferrovanadium was added. In the third heat less oxygen was found in the ferrovanadium-treated ingot, but this ingot had been treated with some aluminum as well. Apparently any oxygen present in these specimens as vanadium oxide was recovered.

TABLE 14.—*Change in apparent oxygen content, as determined by the vacuum-fusion method, produced by addition of aluminum or ferrovanadium in the mold of successive ingots of heats of killed and effervescent steel*

[Al_2O_3 content of the first six specimens shown in last column]

Type of steel	Ingot No.	Mold addition to 3-ton ingot	Structure of carburized layer	Oxygen content O_2	Al_2O_3 content
				Per cent	*Per cent*
Heat 1, 0.11 to 0.14 per cent C, killed steel.	8	4 pounds shot aluminum	Abnormal	[1]	0.025
	9	25 pounds 37 percent ferrovanadium	do	0.003	.003
	10	Regular ingot, no mold addition	Normal	.003	.002
Heat 2, 0.11 to 0.14 per cent C, effervescing steel.	8	4 pounds shot aluminum	Abnormal	[1]	.014
	9	25 pounds 37 per cent ferrovanadium	do	.011	.002
	10	Regular ingot, no mold addition	Normal	.010	.001
Heat 3, 0.11 to 0.14 per cent C, effervescing steel.	8	5 pounds shot aluminum	Abnormal	[1]	----------
	9	25 pounds ferrovanadium, 1½ pounds shot aluminum.	do	.002	----------
	10	Regular ingot, no mold addition	Normal	.009	----------
Heat 4, 0.13 to 0.16 per cent C, ½ per cent Ni, killed steel.	7	25 pounds 37 per cent ferrovanadium	Abnormal	.013	----------
	8	Regular ingot, no mold addition	Normal	.009	----------

[1] Not detected.

Residue analyses [33] for Al_2O_3 by the hydrochloric-acid method described by Oberhoffer and Amman [34] were made of the first six specimens in Table 14; the results are shown in the last column of that table. The aluminum-treated specimens of both the killed and effervescent steel heats contained appreciable amounts of Al_2O_3. However, the oxygen content in the Al_2O_3 in these specimens (about half by weight of the Al_2O_3) is approximately equal to and by no means greater than the general run of oxygen content of the normal specimens shown in Table 13. The conclusion appears warranted that

[33] Analyses by R. J. Kranauer, assistant scientific aide, Bureau of Standards.
[34] Stahl und Eisen, **47**, pp. 1536–1540; 1927.

in commercial carburizing steels there is no direct relation between the total oxygen content and abnormality.

Gat [35] has reported that abnormal steel usually contains much more oxygen than normal steel. This is not borne out by the results at the bureau. Almost all of the normal and abnormal steel specimens analyzed at the bureau had the same range of oxygen content as Gat's normal specimens. His abnormal specimens, however, had unusually high oxygen content. (It should be noted that his analyses were made by identically the same method as that formerly used at the Bureau of Standards, which gives only "apparent" oxygen—not that combined as Al_2O_3, so that his experimental evidence is on just the same basis as that of Table 13.) It is indeed possible that the unusually high oxygen content in Gat's abnormal specimens produced the abnormal structure. However, as pointed out in the preliminary discussion of the causes of abnormality, it would be an error in logic to conclude, because an unusually high oxygen content makes steel abnormal, that all or most commercial abnormal steel has a high oxygen content. As shown in Table 13, only 1 out of 14 representative specimens of commercial abnormal steel was found to have a high apparent oxygen content; the rest had an apparent oxygen content as low as or lower than normal steel.

There is, from all the accumulated evidence, no sharp dividing line between normal and abnormal steel on the sole basis of the content in the steel of any single element. The composition as to carbon, manganese, silicon, phosphorus, sulphur (through its effect in binding manganese), aluminum, and oxygen seems to play an indirect part, the governing factor apparently being not the total amount of any element or combination of elements present, but the form of the element or combination of elements. It is not yet possible to determine exactly the total oxygen content of steels containing aluminum, and, even if this could be done, the question as to how much of the oxygen is combined with each of the elements present in the steel would still remain, as would also the question of the size of the various oxide particles.

5. THE SO-CALLED "IRON-CARBON-OXYGEN EUTECTOID"

Gat [35] has concluded that the ferrite which surrounds the hypereutectoid cementite in the carburized layer of abnormal steel is not true ferrite, but an "iron-carbon-oxygen eutectoid," and that abnormality is due to this eutectoid. The only evidence in his paper of the existence of such a constituent was a micrograph of the carburized layer of abnormal steel, rather deeply etched. The constituent surrounding the cementite had a mottled appearance, and

[35] See footnote 5, p. 427.

this was declared to be the "eutectoid." An indistinct needle scratch across the micrograph was also shown, and it was stated the scratch indicated the eutectoid to be harder than ferrite and of about the same hardness as pearlite. It was pointed out in the discussion [37] that overetched ferrite nearly always has the mottled appearance shown, and that ferrite not well polished may have the same appearance even when not overetched. The unreliability of the scratch test in the case at hand was also pointed out. Gat, in his reply to the discussion, submitted additional micrographs showing the appearance of the "eutectoid." In the first micrograph shown the "eutectoid" seemed to be simply ferrite; in these additional micrographs, on the contrary, the "eutectoid" seemed to be nothing but cementite. It had the characteristic shape of coalesced cementite in low-carbon steel, it polished in relief, and it etched dark with sodium picrate. Gat states that in a very low-carbon steel of 0.02 per cent carbon content there was very little of this constituent, but that there was more in steels of higher carbon content. Naturally, the higher carbon steel would have more cementite.

In polished iron or low-carbon steel some of the grains of ferrite often show a mottled appearance, and one is led to wonder whether these mottled grains are different from the rest of the ferrite, and whether possibly they represent a richer solid solution (as in phosphorous streaks) or, perhaps, even a binary or ternary constituent. The mottled appearance, however, may be due to the polishing. A rough mottled structure in some grains results from hurried polishing in the final stages and is a quite common occurrence. A somewhat different sort of mottled surface may also be obtained because of prolonged polishing. Figure 24 (*a*) shows the mottled appearance (lighter colored patches) produced in a low-carbon steel by long polishing with magnesia in the final stage. A strong impression may be gained that the patches represent another constituent than ferrite or that they are at least appreciably different from the rest of the ferrite. After deeper etching, however, as in Figure 24 (*b*), it seems clear that the mottled patches are not some other constituent but are the result of the covering over by the final polishing of series of intersecting scratches in the ferrite. Under the microscope the connection between the mottled appearance and polishing scratches is well brought out by changing focus. The mottled appearance in ferrite is found not only in abnormal steel, but in normal steel as well. Figure 24 (*c*) and (*d*) shows the mottled appearance of the ferrite in a normal and abnormal specimen. At the bureau, repeated attempts by polishing and etching to reveal any consistent difference in the appearance of the ferrite between normal and abnormal steel have been without result. The "eutectoid" has not been detected

[37] G. F. Comstock, "Discussion of abnormal steels," Trans. Am. Soc. Steel Treat., 12, p. 415; 1927.

either in the carburized layer or in the core of specially prepared high-oxygen iron.[38] Despite the many workers on this problem, no one has been able to show a difference between normal and abnormal steel by means of the copper-bearing etching reagents, such as Stead's, Le Chatelier's, etc. These reagents ought to be very sensitive in revealing the "iron-carbon-oxygen eutectoid," if it exists. It seems justifiable to conclude that the "eutectoid," of itself, and as an explanation of abnormality, is still unproved.

6. INCLUSIONS

In surveys of the inclusions in normal and abnormal steel it was usually possible to detect more alumina inclusions in the abnormal specimens. No very definite relation was found, however, between abnormality and the total number of inclusions. It should be noted that a characteristic of the structure of the carburized layer of abnormal steel is its uniformity; even though variations and segregations of abnormality are frequent, in general, the abnormal features extend very uniformly. It is true that near decided segregations of oxide inclusions the abnormality often becomes most pronounced, but scattered individual inclusions do not seem to exert any noticeable influence on the normality of the surrounding structure. This is one of the reasons why it has been considered that the oxides responsible for abnormality are present not as ordinary sized inclusions but as submicroscopic inclusions which are assumed to be very evenly distributed. From the term "submicroscopic" it is evident that such inclusions can not be detected under the microscope. Normal and abnormal steel specimens carefully polished were examined at high magnification for the very smallest inclusions that can be distinguished to see if there is any difference between normal and abnormal steel in the number of these particles, but no differences were found.

In X-ray diffraction patterns[39] of specimens of commercial normal and abnormal carburizing steels similar in chemical composition the positions of the bands coincided exactly in the two types of steel. The bands in the pattern of the normal steel were somewhat more discontinuous, an indication that it had a coarser grain than the abnormal steel. The only pattern obtained in each case was of the steel; no signs whatever of the pattern of the inclusions can be obtained by this method. It has been reported[40] that abnormal steel specimens showed a greater lattice parameter than normal steel specimens. This conclusion has been questioned,[41] however, on the

[38] C. H. Herty, jr., J. M. Gaines, B. M. Larsen, W. A. Simpkins, The Physical Chemistry of Steel Making: The Solubility of Iron Oxide in Iron, Bulletin 34, 1928. Mining and Metallurgical Investigations, Carnegie Institute of Technology.

[39] Frederick Sillers, jr., junior metallurgist, Bureau of Standards.

[40] O. E. Harder, L. J. Weber, and T. E. Jerabek (see footnote 9, p. 430).

[41] See footnote 9, p. 430; discussion of paper by Frederick Sillers, jr.

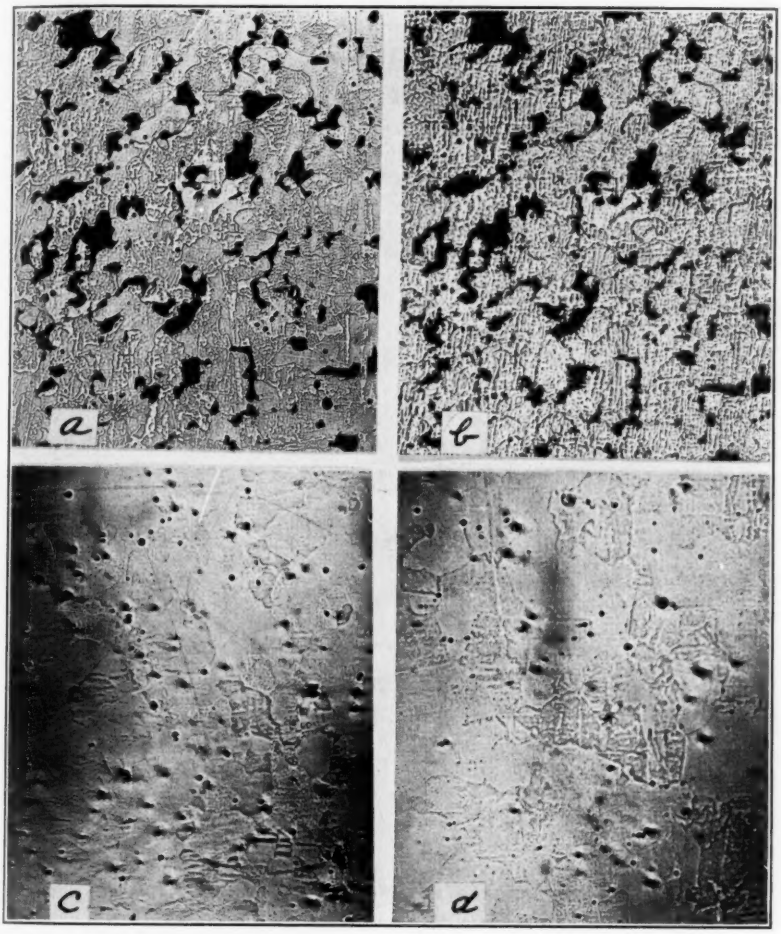

FIG. 24.—*Mottled appearance of ferrite due to polishing.* × *100*

a, after long polishing with magnesia; slightly etched with 5 per cent picric acid in alcohol; the mottled appearance is very marked; *b*, same spot as *a*, more deeply etched. This micrograph suggests very strongly that the mottled appearance is due to scratches; *c*, not etched; mottled appearance after prolonged polishing with magnesia of a normal steel specimen; *d*, not etched; mottled appearance after prolonged polishing with magnesia of an abnormal steel specimen. The mottled structure in *c* and *d* was made more evident by "stopping down" the iris diaphragm

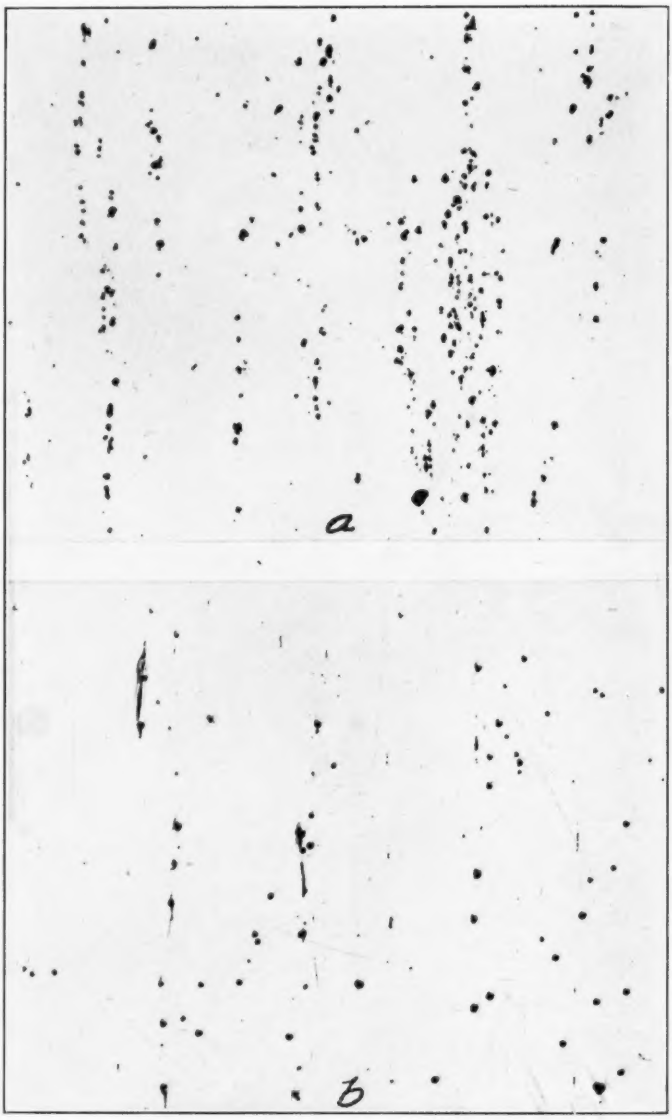

FIG. 25.—*Inclusions in commercial, normal, and abnormal carburizing steel.* × *100*

a, abnormal steel; *b*, normal steel

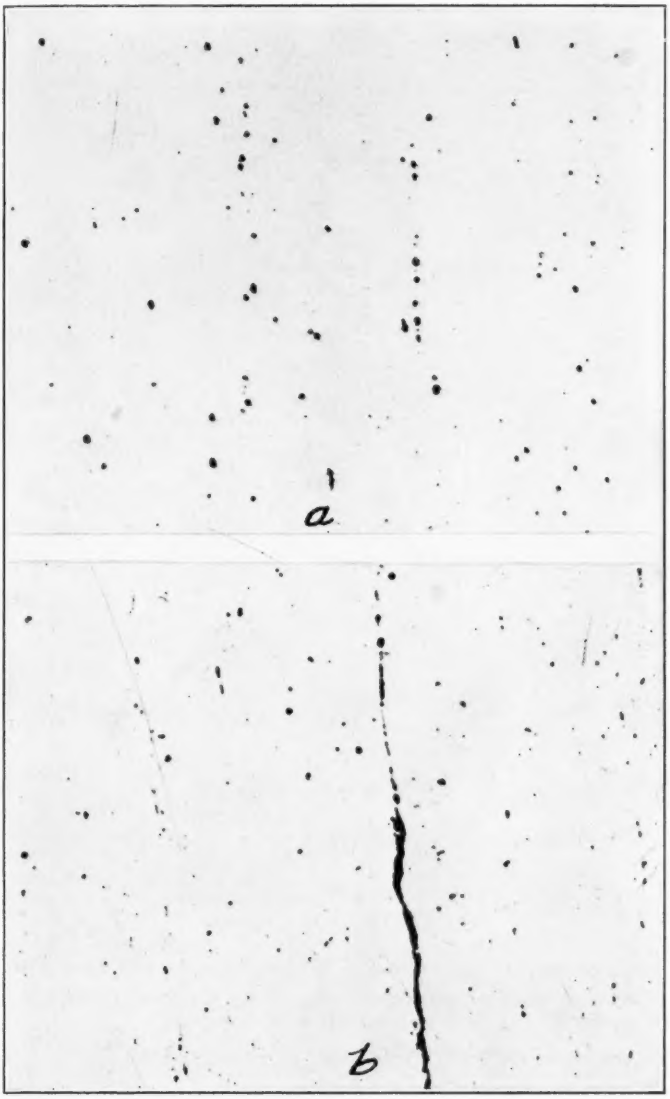

FIG. 26.—*Inclusions in commercial, normal, and abnormal carburizing steel.* × *100*

a, abnormal steel; *b*, normal steel

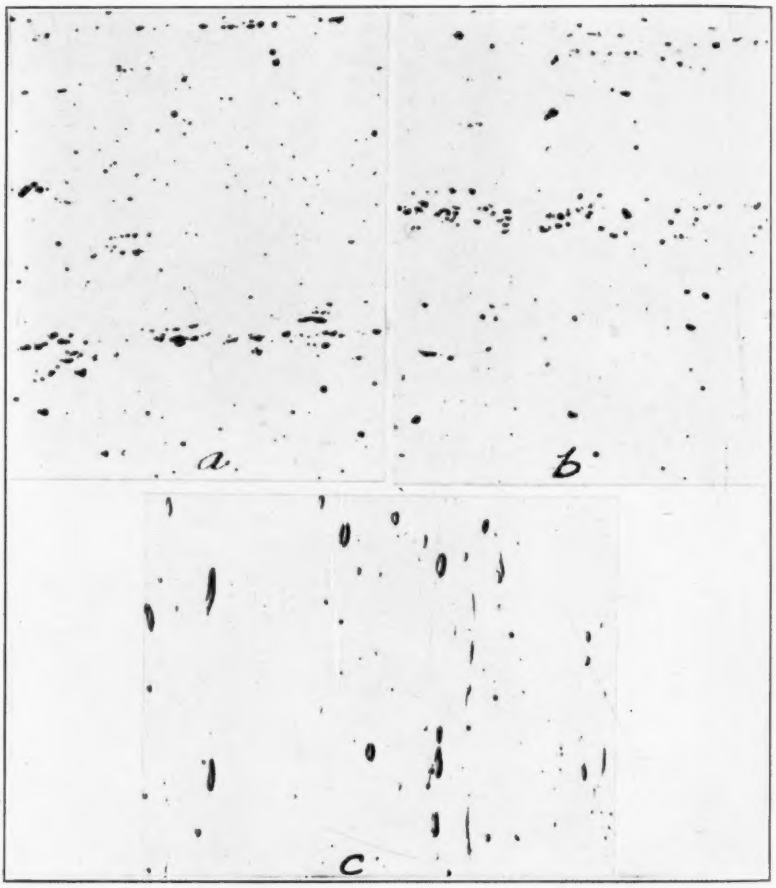

FIG. 27.—*Inclusions in commercial, normal, and abnormal carburizing steel.* × *100*

a, abnormal steel; *b*, normal steel; *c*, special manganese carburizing steel

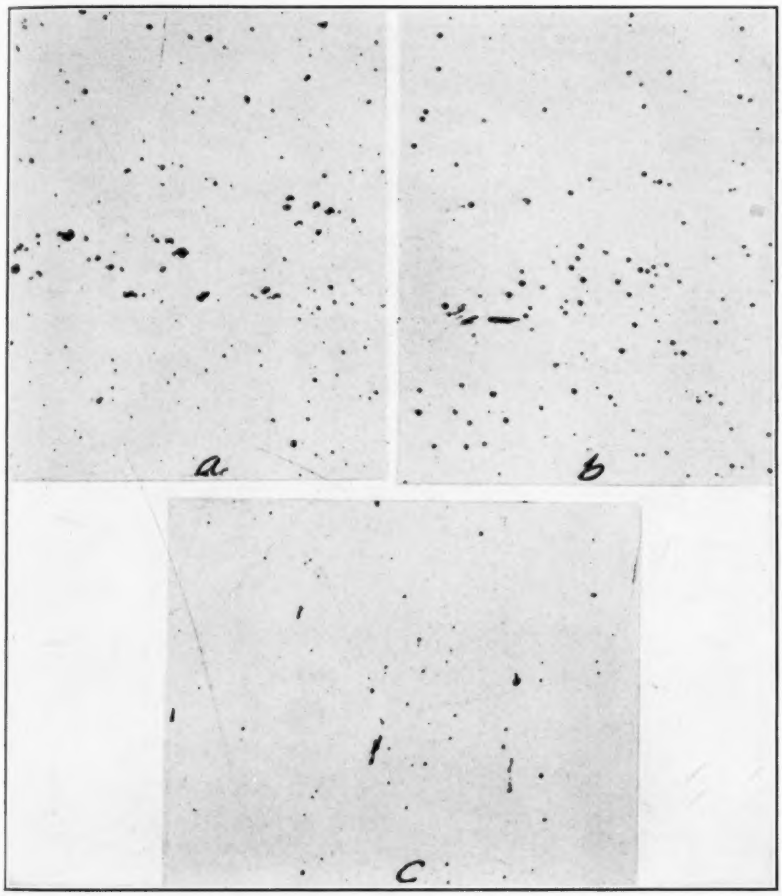

FIG. 28.—*Inclusions in specially treated steel; heat 1, 0.11 to 0.14 per cent carbon killed steel.* × *100*

a, treated with 4 pounds aluminum in the mold; abnormal; *b*, treated with 25 pounds 37 per cent ferrovanadium in the mold; abnormal; *c*, regular untreated ingot from same heat; normal

UNIV. OF MICH.

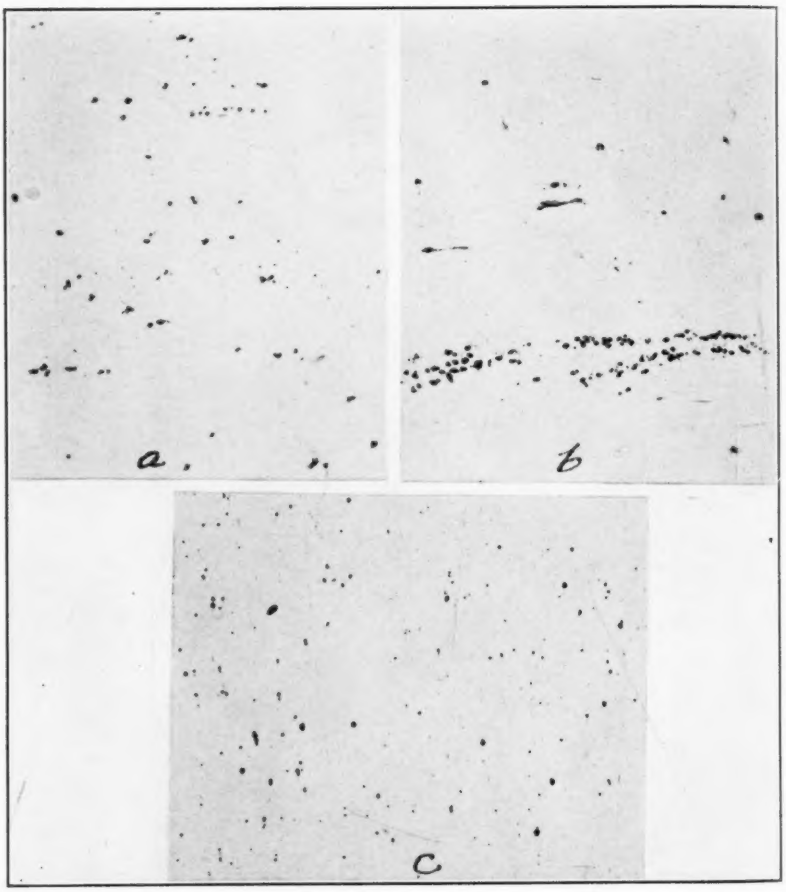

FIG. 29.—*Inclusions in specially treated steel; heat 2, 0.12 per cent carbon effervescent steel.* × *100*

a, treated with 4 pounds aluminum in the mold; abnormal; *b*, treated with 25 pounds 37 per cent ferrovanadium in the mold; abnormal; *c*, regular untreated ingot from the same heat; normal

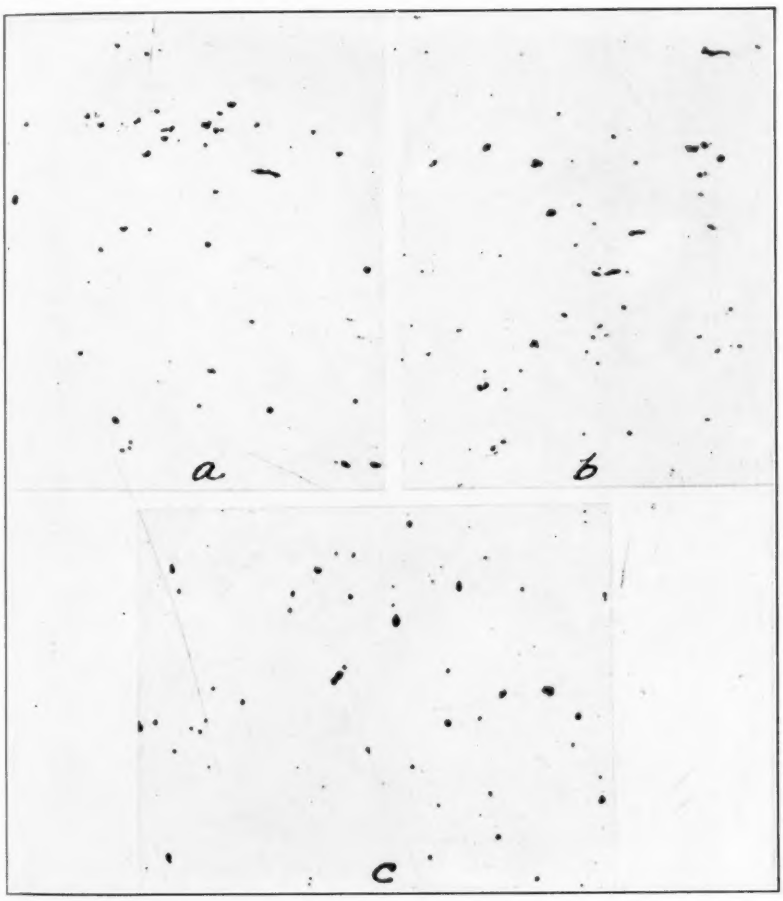

FIG. 30.—*Inclusions in specially treated steel; heat 3, 0.12 per cent carbon effervescent steel.* × *100*

a, treated with 5 pounds of aluminum in the mold; abnormal; *b*, treated with 25 pounds of 37 per cent ferrovanadium and 1½ pounds of aluminum in the mold; abnormal; *c*, regular untreated ingot from the same heat; normal

UNIV. OF MICH.

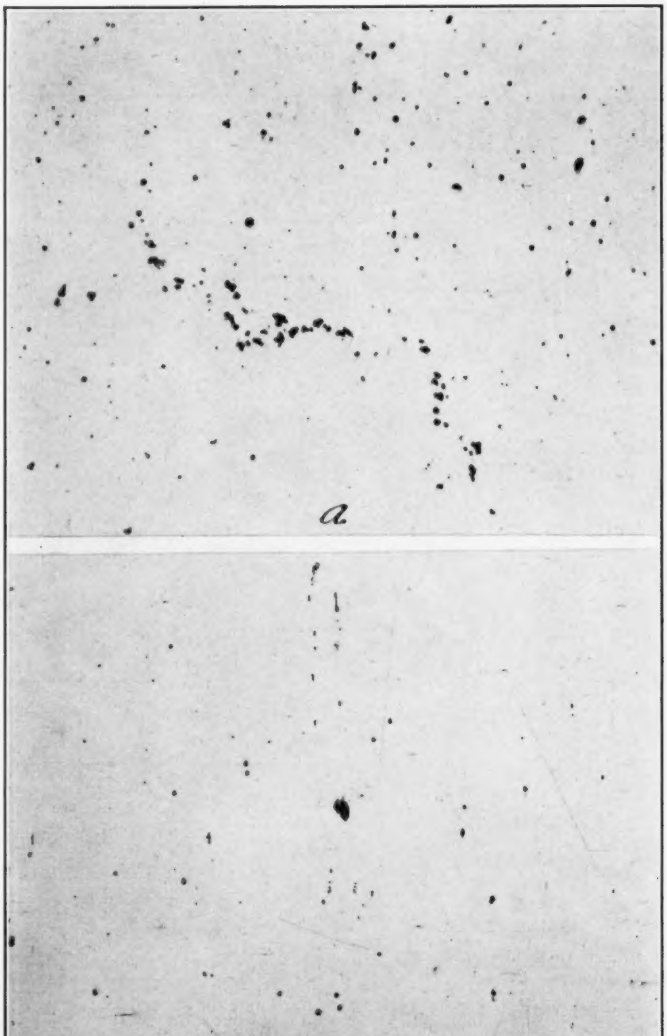

FIG. 31.—*Inclusions in specially treated steel; heat 4, 0.13 to 0.16 per cent carbon, one-half per cent nickel killed steel.* × *100*

a, treated with 25 pounds of 37 per cent ferrovanadium in the mold; abnormal; *b*, regular untreated ingot from same heat; normal

ground that the observed "spread" of the diffraction bands may have been due to distortion of the crystals produced in preparing the specimen in powdered form, or to differences in chemical composition of the specimens used.

Analyses were made by the Dickenson [42] method of the oxide residues of the first six specimens of Table 14. The results are given in Table 15. They are of general interest in connection with the causes of abnormality, although they do not appear to throw any new light on the problem.

TABLE 15.—*Oxide residues by the Dickenson method* [1] *of successive ingots of killed and effervescent steel treated in the mold with aluminum or ferrovanadium, or untreated*

[The specimens are the same as the first six in Table 14 and are numbered according to their order in Table 14]

Specimen No.	Total residue	Al_2O_3	SiO_2	FeO	MnO
	Per cent	*Per cent*	*Per cent*	*Per cent*	*Per cent*
1	0.032	0.022	0.001	0.009	(2)
2	.007	.003	.002	.002	(2)
3	.008	.002	.002	.003	<0.001
4	.028	.017	.001	.010	.001
5	.034	<.001	.010	.011	.01
6	.01	<.001	.001	.008	(2)
	Percentage composition of the total residues				
1	0.032	68	3	29	----------
2	.007	34	31	30	----------
3	.008	21	22	42	6
4	.028	60	3	35	----------
5	.034	<4	28	32	28
6	.01	<10	7	75	----------

[1] Analyses by H. A. Bright, associate chemist, Bureau of Standards. [2] Not detected.

Although the ordinary-sized inclusions may not have a direct bearing on the cause of abnormality, it is still of interest to show representative micrographs of the number of inclusions in normal and abnormal steel. Figures 25, 26, and 27 show the inclusions in representative specimens of commercial normal and abnormal steels. In these micrographs the abnormal specimens can usually, although not always, be recognized by the presence of alumina inclusions. Often, however, normal specimens also contain alumina inclusions. The total number of inclusions in a normal and abnormal specimen may be very nearly the same. An excessive number of alumina inclusions generally indicates abnormal steel. It should be noted (fig. 27) that the special manganese carburizing steel (of high sulphur content), which was normal, contained a large number of inclusions. The presence of manganese sulphide inclusions appears to have no effect whatever on the degree of normality; specimens of screw stock con-

[42] J. H. S. Dickenson, "A note on the distribution of silicates in steel ingots," Iron & Steel Inst., **103**, p. 177; 1926.

taining a large number of these inclusions had perfectly normal structures. Figures 28, 29, 30, and 31 show the inclusions in the steels of Table 12 deoxidized with aluminum or ferrovanadium. The steels treated with aluminum or ferrovanadium in the mold generally, although not always, showed more inclusions than the regular untreated steel from the same heats.

In Figure 29 the number of visible inclusions in the steel treated with aluminum and ferrovanadium in the mold does not seem to be greater than the number in the regular steel from the same heat, although the treated ingot of this heat seemed to be just as abnormal as the treated ingots of the other heats. Such evidence appears to be in line with the supposition of McQuaid and Ehn that abnormality is mainly due to submicroscopic inclusions, presumably oxides. As already stated, such inclusions are not revealed by the microscope or by the X-ray spectograph. Various methods of chemical analysis, in which nonmetallic residues are obtained after solution of the steel in some reagent, may be considered as a means of detecting the presence of such inclusions, but the question at once arises whether or not a reagent which has little or no solvent action on a large nonmetallic inclusion of a given composition will be similarly inert when the inclusion is of colloidal dimensions. In any method involving filtrations it is uncertain that all or any of the finely divided material will be retained by the filter. In decantation losses may arise through the failure of colloidal-sized particles to settle. Ignition of the residue is likely to change the state of oxidation of oxides of metals with two or more valencies.

Unless some new method or research tool becomes available, it appears that our understanding of the causes of abnormality must continue to rest on circumstantial evidence. The method of attack which should give most information of direct use in the control of normality appears to be that of synthesis rather than analysis; that is, work along the general lines of the deoxidation experiments made in the cooperative work at the Central Alloy Steel Corporation's plant. By the choice of raw materials and the control of furnace operation, slag composition, and the time and method of adding deoxidizers, several steel plants have the matter sufficiently under control so that they feel able to produce at will steel of a predetermined degree of normality and abnormality. Comprehensive laboratory work along these lines, in which the different variables may be kept under better control than in commercial furnaces, is in progress in the Bureau of Mines laboratories in Pittsburgh. Work on the solubility of various oxides in liquid and solid steel at different temperatures, the chemical action of different slags, the composition of the inclusions formed under different conditions, the rate of reaction when different deoxidizers are added to the steel in various stages of the "finishing" process,

and on the gravity elimination of inclusions may be expected to throw light on this problem. When the relation of these factors to abnormality are studied, the cause and control of abnormality will no doubt be better understood, but even then the evidence would still be circumstantial. However, the slip-interference theory of hardening by precipitation and agglomeration to a critical submicroscopic size of particles of metallic compounds, which so satisfactorily explains the age-hardening of duralumin, likewise rests upon evidence which is essentially circumstantial, and yet that fact has been no bar to the general acceptance of the theory.

In regard to the mechanism of formation of the abnormal structure, it is evident that, in the absence of any data as to the state in which the oxides presumed to be the cause of abnormality are present, nothing very definite can be said. The fact that the abnormal microstructure possesses two distinct, although related, features—namely, fine grain and coalescence of the cementite—would suggest that the cause of abnormality may be of a duplex nature. Thus, it is, perhaps, possible that the fine grain is due to relatively larger oxide particles (the behavior of foreign particles in obstructing grain growth is, of course, well recognized), whereas the coalescence of the cementite is due to finer particles, or perhaps to dissolved oxides. From the deoxidation tests with aluminum and ferrovanadium (VI, 2) we may, perhaps, infer that the bulk of the material causing abnormality is in a fine state of distribution rather than in solution, since it is rather improbable that aluminum oxide is in true solution in the steel.

7. THERMAL ANALYSES

Differences between normal and abnormal steel have been brought out by thermal analyses [43] (inverse-rate curves) made in the modified Rosenhain furnace in use at the bureau. The differences appeared in the cooling curves of uncarburized samples, but not in the heating curves; they did not appear when the samples were carburized, probably because the modification in the heat effect produced by abnormality is obscured in the presence of high carbon content. It was thought at first that thermal analysis might possibly be used as a supplementary method for detecting abnormal steel, but this idea had to be given up, since the differences in the curves between normal and abnormal steel are rather slight and are readily obscured by differences in chemical composition. The significant features of the thermal analysis curves are best illustrated by a series of cooling curves of the steels of Table 12 treated with aluminum or ferrovanadium in the mold, shown in Figure 32. The same features were brought out in thermal analyses of a number of samples of commercial normal and abnormal steel.

[43] W. A. Tucker, assistant scientific aid, Bureau of Standards.

As may be seen in Figure 32, the Ar_1 transformation in the abnormal specimen of each heat occurred at a slightly higher temperature than that of the corresponding normal steel. The Ar_1 transformation was also sharper; that is, it occurred more nearly at one temperature, and the maximum time of arrest at the Ar_1 point was greater in the abnormal than in the normal steels. Thus, in heat 1, in the aluminum-treated steel, the maximum time of arrest at Ar_1 was 21 seconds; in the regular ingot it was 14 seconds. As can be seen in Figure 32, in the rest of the heats there was a corresponding difference between the normal and abnormal steels in the maximum time of arrest at Ar_1.

A possible interpretation of the above-mentioned characteristics of the cooling curves of abnormal steel is that they are a reflection of the same conditions as produce coalescence and separation of the cementite from the ferrite in the carburized layer of abnormal steel. The fact that the Ar_1 point occurs at a higher temperature and more sharply at a single temperature in abnormal than in normal steel would indicate that the cementite forms more readily when the Ar_1 temperature is reached in abnormal than in normal steel. This is the probable reason why abnormal steel is more prone to give soft spots. Cementite forms more readily in abnormal steel, and so, on quenching, troostite is more readily produced and a faster cooling rate is required to retain martensite. Harder [44] has given a discussion of the mechanism of formation of the abnormal structure in the carburized layer, based largely on deductions from the microstructure, that in abnormal steel cementite is less soluble in the austenite and has represented this by lines on the iron-carbon equilibrium diagram. It is very doubtful, however, whether the existing experimental data justify such a representation. The thermal analyses made in this work indicated that, upon heating, the cementite of abnormal steel goes into solution at the same rate as in normal steel, the heating curves of the two kinds of steel (uncarburized stock) being similar. The difference between the two was shown only on cooling; apparently the cementite in abnormal steel forms more abruptly, perhaps because of the presence of submicroscopic particles of oxides acting as nuclei for precipitation, as stated in Ehn's original theory.

VII. PRACTICAL ASPECTS

The problem of soft spots in casehardened articles due to the character of the steel used presents no difficulties that can not be overcome by increasing the cooling rate in quenching or by changing the type of steel used. The real question is how to deal with the problem in any particular plant. This will depend on considerations

[44] See footnote 9, p. 430.

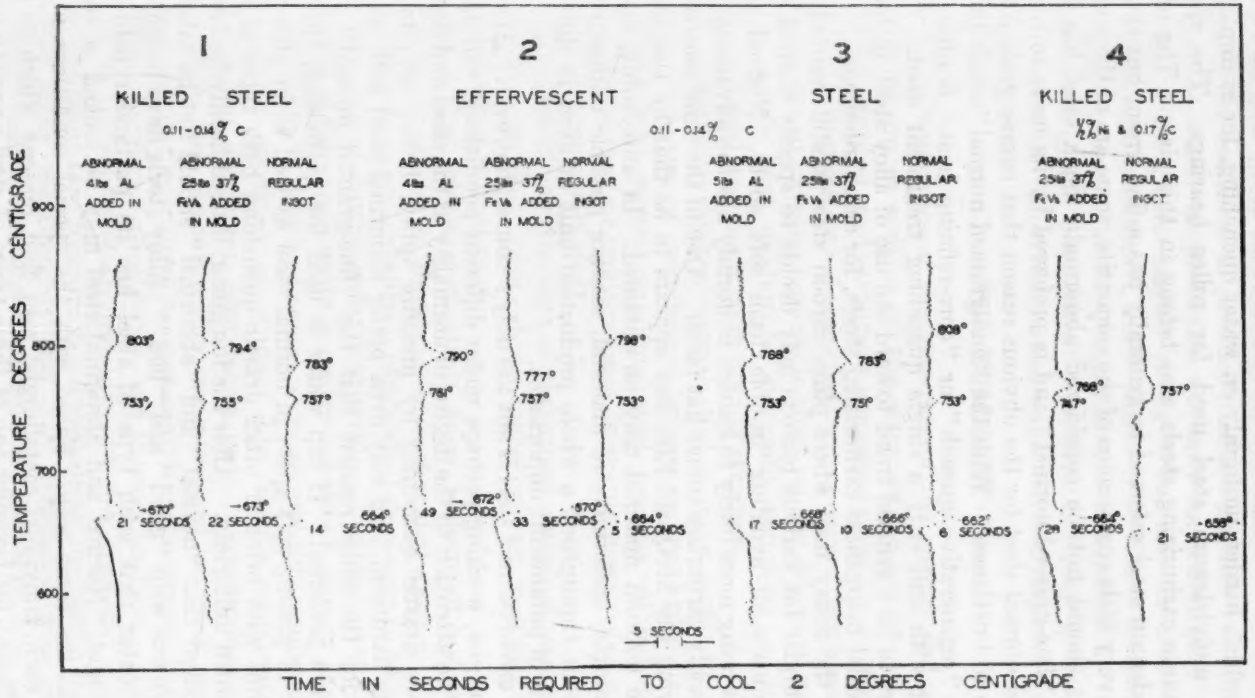

FIG. 32.—*Inverse-rate cooling curves of normal and abnormal steel*

The Ar_1 point in the abnormal specimens occurred at a slightly higher temperature and more sharply at one temperature than in the normal specimens

of practice, equipment, quality, cost, and the personal preference of the metallurgist, as with any practical metallurgical problem.

One solution of the problem has been the use of the alloy carburizing steels which harden uniformly on water quenching, for example, the nickel molybdenum steel used for roller bearings. The special manganese carburizing steels also belong in this class. The nickel molybdenum steel referred to generally has a fine grain, but since it shows very little coalescence of the cementite, some are of the opinion that it should not be considered abnormal. Such steel has been called "fine-grained normal" and is preferred by its users to coarse-grained normal steel for the obvious reason that coarse grain is conducive to brittleness. With the "fine-grained normal" steels the so-called "regenerative quench" or "core-refining heat" is often dispensed with and only a single quenching treatment used. There appears to be a general trend toward the use of alloy steel, including the special manganese carburizing steels, for casehardening.

For the many uses where plain carbon steel is still wanted, the metallurgist for various reasons may decide to specify normal steel as a means of avoiding trouble from soft spots. Normal steel, besides being more likely to harden uniformly has the advantage that it generally carburizes somewhat faster. One of the chief reasons for specifying the McQuaid-Ehn test appears to be that by this means a more uniform material may be obtained. In automobile plants, for example, uniformity of material, making possible uniformity of treatment throughout a whole production unit or even in different units, is of paramount importance.

The use of normal steel is not the only recourse, however. Another metallurgist, working perhaps under different conditions, will decide to pay no attention to the degree of normality of the steel and depend on more drastic quenching by pressure spray, brine, etc., to get uniform hardness. He may even prefer abnormal steel with its fine grain, for the same reasons that the "fine-grained normal" alloy steels are preferred. If tap water is used for quenching, there is danger of getting soft spots in normal steel as well as in abnormal steel, but with brine or other drastic quenching both types of steel will harden uniformly. After the foregoing, it need hardly be further emphasized that "normal" and "abnormal" should not be taken as synonymous with "good" and "bad." Alloy steels have taught the steel treater that each type of steel has its particular field and advantages. Normal and abnormal steel may be looked at somewhat in the same way. Faced with the special conditions in his plant, each individual metallurgist may find reasons which to him are sufficient for favoring one of these types of steel over the other.

From the standpoint of the steel maker the results of the deoxidation tests described indicate that, after suitable experimentation, a

steel manufacturer ought to be able to produce normal and abnormal steel at will, subject, of course, to the limitations of chemical composition. Except under very unusual circumstances, it would hardly be possible, for instance, to produce a coarse-grained vanadium steel. Fine grain would be preferable not only for carburizing steel, on account of the greater impact strength, but also for other uses, as, for instance, in sheet steel where coarse-grain size is detrimental for certain purposes. The McQuaid-Ehn test is sensitive to variations in the melting and deoxidation procedure and may become a very useful test in the hands of the steel maker for studying and controlling his product.

VIII. SUMMARY

The characteristics of the extremes of the normal and abnormal microstructure, as well as of the intermediate gradations, are described. Normal steel carburizes somewhat more deeply than abnormal steel. Segregations of abnormality may occur. Mechanical treatment seems to have no noticeable effect on the degree of normality. After certain heat treatments, especially annealing in hydrogen, abnormal steel assumes a more normal structure.

On water quenching abnormal steel is more prone to give soft spots. However, on quenching in brine or sodium hydroxide solution abnormal steel, as well as normal steel, hardens without soft spots. It was found that air or other gases ordinarily dissolved in quenching water may cause soft spots.

After carburizing, the core of carburized abnormal steel, probably because of its finer grain, shows higher impact strength than the corresponding normal steel.

High-carbon steels (tool steels) show features of normality and abnormality similar to those met in carburizing steels.

Most cases of abnormality in commercial steel seem to be associated with the use of aluminum for deoxidizing. Nothing arose during the progress of the experiments to controvert Ehn's theory that abnormality is due to dissolved oxides or, more probably, undissolved submicroscopic particles of oxides. The results of spectrographic analyses for aluminum, deoxidation tests of commercial steel with aluminum and ferrovanadium, examinations of small test ingots taken during the progress of open-hearth heats, oxygen and alumina analyses, surveys for inclusions, and thermal analyses—made in the study of the causes of abnormality—are described and discussed.

The authors take pleasure in acknowledging the hearty cooperation of the many metallurgists and steel companies who generously provided samples for the investigation, and who also gave their advice and freely divulged the results of their experience on this problem, making the study a truly cooperative one between the bureau and the industry.

It would make too extended a list to credit everyone who has shown an active interest in the work, but, in addition to those who have been mentioned in the footnotes, acknowledgment is due the following: Bureau of Mines, Crucible Steel Co., Driver Harris Co., Jones & Laughlin Co., Chrysler Corporation, Pittsburgh Crucible Steel Co., Reo Motor Co., and Timken Roller Bearing Co. The passing mention in the footnotes really does not do justice to the amount of time spent on the work at the Central Alloy Steel Corporation and the Trumbull Steel Co.; in the experiments made at the latter plant the Titanium Alloy Manufacturing Co. and the Union Carbide & Carbon Co. aided materially. E. W. Upham, of the Chrysler Corporation, was largely instrumental in the bureau undertaking the study of the problem.

WASHINGTON, May 23, 1928.

STRAIN MARKINGS IN MILD STEEL UNDER TENSION

By Henry S. Rawdon

ABSTRACT

Polished tension bars were examined during the tension test to determine the stress and corresponding elongation at which strain markings appear. The mode of propagation of the strain effect throughout the bar has been determined by repeated loadings on the same bar. Hardness increases about 5 per cent during the formation of strain markings, this increase being about one-tenth of that occurring when the same steel is stressed to fracture. Strain markings result from deformation by slip within a relatively few suitably oriented ferrite grains. This slip is accompanied by a general shifting and tilting of neighboring grains, however, which results in the characteristic rippled surface appearance of strain markings on steel. Similar markings were observed on an aluminum alloy.

CONTENTS

	Page
I. Introduction	467
II. Materials and method	469
III. Results:	
1. Tension tests of polished specimens	470
2. Effect of repeated loading	472
3. Hardness of strained areas	474
4. X-ray examination	476
5. Microstructure	477
6. Comparison with an aluminum alloy	479
IV. Discussion	479
V. Summary	484

I. INTRODUCTION

The mode of deformation of metals under stress has long been a subject of interest. The observation of Lüders, whose name is generally associated with the macroscopic markings on metals which have been strained, dating from about 1854, was one of the earliest published on this subject.[1] The same phenomena had been observed, however, as early as 1836 by Piobert.[2] The name of Hartmann, who, at a much later date,[3] studied this phenomenon at great length, is

[1] W. Lüders, Ueber die Aeusserung der Elasticität an Stahlartigen Eisenstaben und Stahlstaben, und über eine beim Biegen solcher Stabe beobachtete Molecularbewegung, Dingler's Pol. J., **155**, p. 18; 1860.

[2] Memorial de l'Artillerie, **V**,, p. 505; 1842; see, also, C. Fremont, La cause de la formation du palier dans l'essai de traction des aciers doux, La Génie Civil, **82**, p. 176; 1923.

[3] L. Hartmann, Sur la distribution des déformations dans les métaux soumis à des efforts, Comptes Rendus, **118**, p. 520; 1894; Distribution des déformations dans les métaux soumis à des efforts (booklet); 1896, Berger-Levrault et Cie., Paris and Nancy; Phénomènes qui accompagnent la déformation permanente des métaux, Congrès International des Méthodes d'essai, **1**, p. 95; 1901; Sur le mécanisme de la déformation permanente dans les métaux soumis à l'extension, Comptes Rendus, **152**, pp. 1005, 1084, 1233; 1911.

often linked with that of Lüders in referring to the markings on steel specimens which have been permanently strained.

Numerous other descriptions of this phenomenon have appeared in the technical literature. Of these, the articles by Gulliver [4] and Mason [5] are especially noteworthy, since these investigators have done much in correlating the formation of the strain lines with the magnitude and nature of the stress applied to the specimen.

Aside from the extensive studies which have been made on the microscopic aspects of the deformation of metals, the most recent work dealings with the general subject of strain in metals relates primarily to methods for revealing the strained portions. The special etching method described by Fry [6] for revealing strain lines and areas in the macrostructure of steel and applied later by Turner and Jevons [7] and by Fell [8] in their very comprehensive studies has proved a noteworthy contribution to our knowledge on this subject.

An excellent series of descriptive articles on the general subject of strain lines is that of Seigel.[9] The most recent article on this subject, which deals particularly with the theoretical aspects of the underlying cause, is that of Takaba and Okuda.[10]

Despite the numerous articles which have been published on the subject of strain lines there are a number of very important points which are still obscure. Although it is generally assumed that the formation of these lines is a special case of the deformation of a metal by slip, very little has been accomplished to make clear the microstructural changes which underlie this phenomenon. In addition to the stress at which the strain lines first appear, it is also of importance to know the behavior of the material, as revealed by the change in the number and extent of the strain lines; that is, their "propagation" throughout the stressed specimen, as the specimen is maintained under constant load. Considerable information on this "propagation of strain" has been given by the work of Turner and Jevons,[11] who applied the etching method of Fry to a series of tension bars which had been permanently strained, each by a different amount. A series of observations taken on the same bar while under stress, however, would be preferable in many respects to the use of a number

[4] G. H. Gulliver, "Some phenomena of permanent deformation in metals," Proc. Inst. Mech. Engrs., pts. 1 and 2, p. 141; 1905.

[5] W. Mason, "The Lüder's lines on mild steel," Proc. Phys. Soc. of London, **23**, p. 305; 1911.

[6] Adolf Fry, "Kraftwirkungsfiguren in Flusseisen dargestellt durch ein neues Ätzverfahren," Kruppsche Monatshefte, **2**, p. 117, 1921; Stahl und Eisen, **41**, p. 1093; 1921.

[7] T. H. Turner and J. D. Jevons, "The detection of strain in mild steel," J. Iron Steel Inst., **111**, p. 169; 1925. J. D. Jevons, "Strain detection in mild steel by special etching," J. Iron Steel Inst., **111**, p. 191; 1925.

[8] Eric W. Fell, "Strain in steel (experiments upon the nature and detection of certain phenomena observed in permanently deformed steel)," J. Iron Steel Inst., Carnegie Schol. Mem., **16**, p. 101; 1927.

[9] J. Seigel, "Les premièrs déformations permanentes dans les aciers doux," Le Génie Civil, **88**, pp. 315, 332, 357; 1926; also "Condition d'apparition des lignes de cession," Le Génie Civil, **90**, 576; 1927.

[10] Ititaro Takaba and Katumi Okuda, "Die wichtigsten Eigenschaften und die Theorie der Fliessfiguren," Archiv für das Eisenhüttenwesen, **1**, No. 7, p. 511; January, 1928.

[11] See footnote 7, p. 468.

of different bars of the same nominal composition. The present study has been carried out in the endeavor to obtain more definite information on these points. The work has been restricted almost entirely to low-carbon steel bars stressed in tension; a few supplementary observations, however, were made on an aluminum alloy by way of comparison.

II. MATERIALS AND METHOD

Low-carbon steel of boiler-plate grade was used for most of the specimens. The average composition of this steel, which was used in the hot-rolled condition, was: C, 0.17 per cent; Mn, 40; P, 0.02; S. 0.033. For the X-ray study, as well as for some of the microscopic examinations, steel strip, 0.0025 inch thick, made by the continuous rolling process was used. This strip steel contained 0.23 per cent carbon and 0.35 per cent manganese, and prior to use the material was annealed two hours at 750° C., a number of strips of the thin material being clamped between two steel plates in order to reduce oxidation and to keep the strips flat.

Tension specimens, 18 inches in length and 2 inches wide, with a reduced section 1½ inches wide, the parallel portion of which was 9 inches long, were cut from the ⅜-inch boiler plate. On one side of the bar the oxide scale was allowed to remain intact while the other side was machined free from scale. The reduced section of the bar was polished before testing. A few specimens of boiler-plate steel, with dimensions one-half those given, were also used. All of the specimens cut from the thin strip steel were of the smaller size.

The specimens were stressed in an Amsler testing machine, three types of grips for holding the specimens being used in different tests. All of the 18-inch bars were tested by means of the usual type of wedge grips. For most of the smaller specimens grips designed to give a very close approach to true axial tension [12] were used. A special form of grip,[13] which has been used with pronounced success at the Bureau of Standards for the tension testing of small members which have been formed out of thin sheet metal, such as the duralumin lattice bars of airship girders, such members being tested in their finished shape—that is, without any preliminary flattening before testing—were also used. The appearance of the strain lines obtained was not essentially different for the three forms of grips; that is, so far as the general character and orientation of the first lines which formed are concerned. The results given below may be considered as representative of results obtained with all of the grips used.

Measurements of the elongation under load were made on a number of the 18-inch bars, a Ewing extensometer being used.

[12] Tentative methods of tension testing of metallic materials, E8-25T, Proc. Am. Soc. Test. Matls., 25, p. 858; 1925.

[13] Developed at the Bureau of Standards by Dr. L. B. Tuckerman, engineer physicist. A description of this device has not been published as yet.

III. RESULTS

1. TENSION TESTS OF POLISHED SPECIMENS

Observations made on the polished surface of a bar as the load was applied, as compared with observations on the other side of the bar from which the oxide (mill scale) had not been removed, showed very plainly that accurate results can be obtained only with a polished specimen. Not only were the sharp lines and fine details of the pattern entirely lacking on the scaled surface, but an appreciably longer time, when the load was maintained constant, was required in order to cause flaking of the oxide and thus reveal the strain markings. When the load was maintained as close as practicable to that which caused strain lines to appear on the polished surface, approximately 30 minutes elapsed before any evidence whatever of flaking of the oxide scale was observed. When the load on a specimen was maintained for a rather long time at a value just above that which was sufficient to cause the formation of strain lines, the effect gradually progressed from the fillet end of the reduced section of the bar toward the center of the specimen. In most cases the effects at the two ends of the bar were essentially the same. As seen on the scaled side of the specimen, the boundary surface between the strained and the unstrained portion as it progressed along the stressed bar "lagged" noticeably behind that as seen on the polished side. On the large tension specimens (18-inch) this "lag" was approximately ½ inch. Observations on strain markings on mild-steel specimens stressed in tension as revealed by the flaking of a coating of mill scale on such material indicate that this method and possibly also methods depending upon the use of other coatings, such as "whitewash,"[14] are most useful in a qualitative sense for locating the strained areas rather than for an accurate indication of the stress necessary for the formation of such strain markings.

When a specimen is stressed in tension, the strain lines first appear in the reduced section just below the fillet. As can be shown by optical methods,[15] in this portion of the stressed bar the stress is considerably greater than elsewhere throughout the reduced section. The lines originate at the edge and progress across the width of the bar. In the specimens used the lines at the two ends originated at opposite sides of the bar. The course of the first one at each end was approximately perpendicular to the axis of the bar instead of following an angle of 45 or 50° as is usually considered typical for such markings. Presumably this was on account of the stress on specimen not being perfectly axial. This effect was noted, however,

[14] Bureau of Standards Miscellaneous Publications No. 72; also R. S. Johnston, Compressive Strength of Column Web Plates and Wide Web Columns, Tech. Paper No. 327, 1926.

[15] E. G. Coker, "Photo-elastic methods of testing." Report of British Asso. Advancement of Science, 92, p. 313, 1924; also, "Elasticity and plasticity," Thirteenth Thomas Hawkley Lecture, Proc. Inst. Mech. Eng., pt. 2, p. 897, 1926.

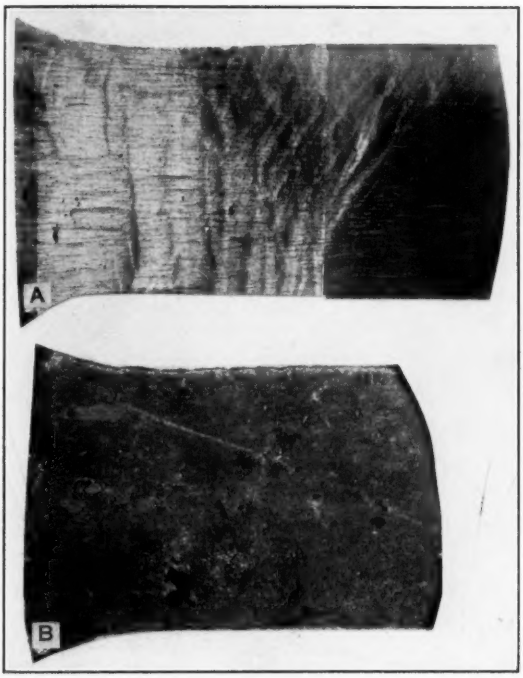

FIG. 1.—*Strain markings on soft steel as seen on a polished surface and on one covered with mill scale.* × 1

Specimen 3 (Table 1), the first line appeared near the fillet at 32,000 lbs./in.², stress was raised to 33,800 lbs./in.² and held five minutes; *a*, the surface was polished before stressing; *b*, opposite surface on which the mill scale was left intact

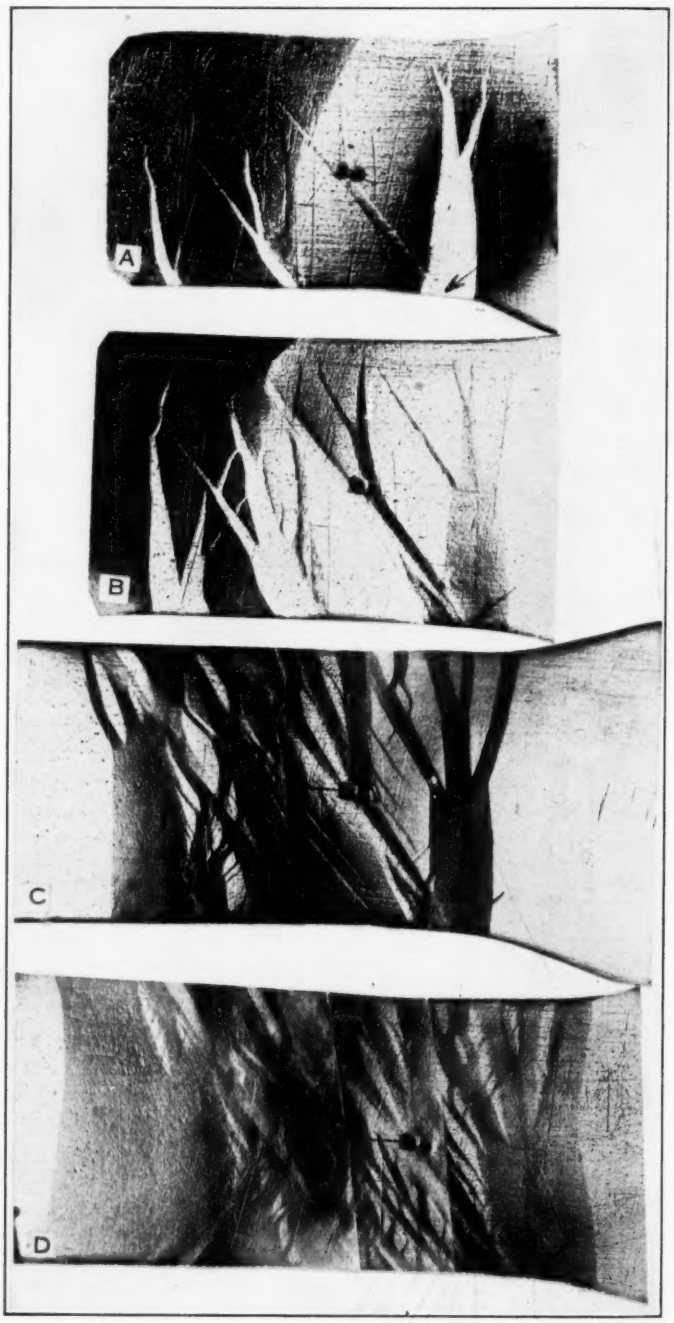

FIG. 3.—*Strain markings which formed at one end of specimen 7 when subjected to the first four loadings of Figure 2.* × *1*

The specimens were photographed under "vertical" illumination; *a*, *b*, *c*, and *d*, loadings 1, 2, 3, and 4. Note that the extensometer readings given in Figure 2 correspond to only those strain markings *within* the gauge marks, in both Figures 3 and 4. The arrows indicate the base of the fillet

on all specimens, regardless of the type of grips used for holding them. In the specimens which were stressed until fracture occurred the location of the fracture appeared to bear no obvious relation to the preexisting strain lines or the point where the "deformation waves" progressing from the two ends met, except in one case. The first strain lines to appear in this specimen (No. 11, Table 1) formed at a considerable distance from the fillet, and the fracture, when the specimen was broken, occurred close to this point.

TABLE 1.—*Summary of specimens of boiler-plate steel tested*

Specimen No. and size	Stress at which strain lines appeared	Length of time load was held	Observations on oxide scale	Disposition made of specimen	Remarks
	Lbs./in.[2]				
1 (18-inch)	33, 800	60 minutes			Strain lines formed just below fillet and progressed toward center; bar was loaded to failure; location of the fracture bore no obvious relation to the strain lines.
2 (18-inch)	[1] 30, 200 33, 800	5 hours	First lines on oxide seen 25 minutes after those on polished surface.	Tension test, ult. ten. str. 56,700 lb./in.[2]; elongation (8 inches), 33 per cent.	
3 (18-inch)	32, 000	5 minutes at 33,800 lbs./in.[2]		Hardness survey after first loading and after breaking, ult. ten. str. 55,600 lbs./in.[2]; elongation (8 inches) 33 per cent.	
4 (18-inch)	33, 400				Increase in Brinell hardness in strain markings, 3 per cent; increase in Rockwell in strain markings, 4.3 per cent; increase in B. H. N. after tension test, 44 per cent; increase in Rockwell after tension test, 43 per cent.
5 (18-inch)	[1] 28, 600 30, 200 32, 000	30 minutes	No flaking of oxide scale.	Examination of microstructure.	
6 (9-inch)	[1] 32, 800 33, 600 34, 300				
7 (18-inch)	[2] 28, 700 29, 700 28, 800 31, 300 30, 600			Tension test, ult. ten. str. 54,000 lbs./in.[2]; elongation (8 inches), 33.5 per cent.	See Figure 2.
8 (9-inch)	33, 500	40 minutes		Used for hardness tests.	Increase in Brinell hardness in strain markings, 3 and 5.5 per cent
9 (9-inch)	33, 300	45 minutes			
10 (18-inch)	[1] 27, 600 31, 000	10 minutes		Used for hardness tests.	Increase in hardness in strain markings, Brinell hardness, 4.5 and 5.5 per cent; Rockwell, 8 per cent. (See fig 7.)
11 (18-inch)	28, 200			Tension test and hardness, ult. ten. str. 56,900 lbs./in.[2]; elongation (8 inches), 31 per cent.	Increase in hardness, Brinell hardness, 47 per cent; Rockwell, 47 per cent. (See fig. 7.)

[1] The stresses given mark the first appearance and successive distinct stages in the development of the strain lines.

[2] The specimen was loaded successively several times, each stress indicating the appearance of new lines for each particular loading.

In Table 1 is given a summary of the results obtained with the specimens of the boiler-plate steel used in the study of strain markings.

2. EFFECT OF REPEATED LOADING

The general character of strain lines and their mode of "propagation" along the length of the bar as it is stressed in tension are best shown by the results obtained with specimen No. 7 (figs. 2 to 6, inclusive), which was loaded five times in succession, the specimen being removed from the machine after each loading for observation and photographing. In order to secure good contrast between the strained areas and the unstrained portions, in the photographing of the specimens, which was done under "vertical" illumination, only a rather small part of the bar could be photographed at a time. The smaller views were then assembled in proper order so as to give the complete picture of the specimen. The diagrams of Figure 6, showing the "growth" in the strained areas with prolonged loading, were obtained by tracing directly on tracing linen the image of the stressed bar, after removal from the machine, as projected on to the ground glass of the metallurgical camera.

The behavior of the specimen when loaded several times and the load released between successive loadings is summarized graphically in Figure 2. The elongation of the specimen was measured on an 8-inch gauge length by means of a Ewing extensometer (18-inch tension bar). Other details, such as the length of time the load was maintained constant, together with the extension produced, the length of time which elapsed between loadings, the stress at which strain markings first appeared during each loading, as well as the different "permanent sets" produced have also been given. It will be noted that in each case the first appearance of strain markings either coincided with the "break" in the stress-strain curve or occurred immediately after the break. When a specimen has been sufficiently strained in a previous loading, so as to produce strain markings, it is exceedingly difficult to detect first indication of the formation of additional strain lines during a subsequent loading.

Figures 3 and 4 clearly show that the strain lines under long continued and successive loadings "grow" by the widening and branching of the preexisting lines as well as by the formation of new lines. In a number of cases (fig. 6, fourth and fifth loading) the "front" of the strained area which extended completely across the specimen was seen to progress uniformly along the length of the bar, from each end, no new or additional strain lines being formed. It is of interest to note that at the end of the fifth loading, when the load was released, these two "fronts" which had not yet met differed appreciably in the angle which they formed with the axis of the specimen; that is,

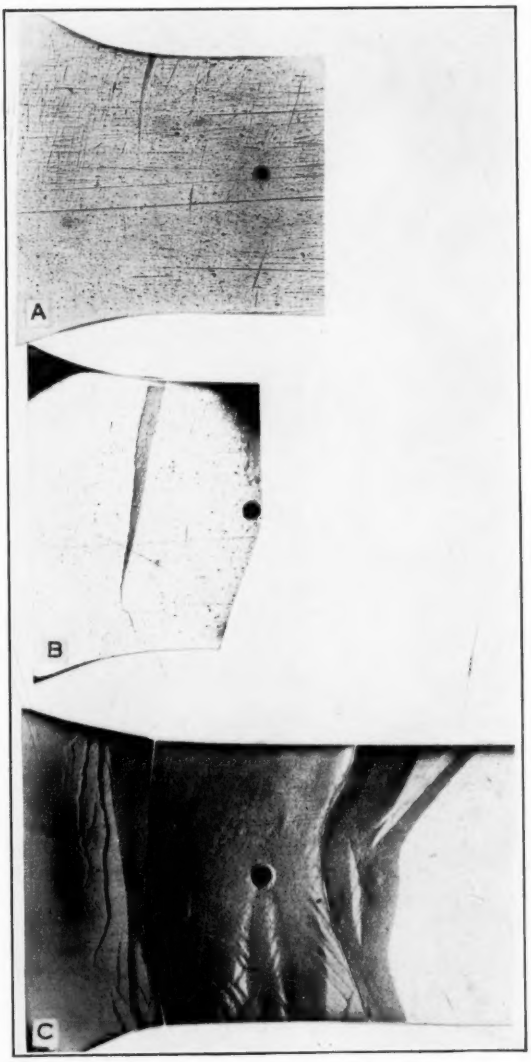

FIG. 4.—*Strain markings which formed at the opposite end of specimen 7, to that shown in Figure 3, when stressed as in Figure 2.* × 1

a, *b*, and *c*, loadings 1, 3, and 4

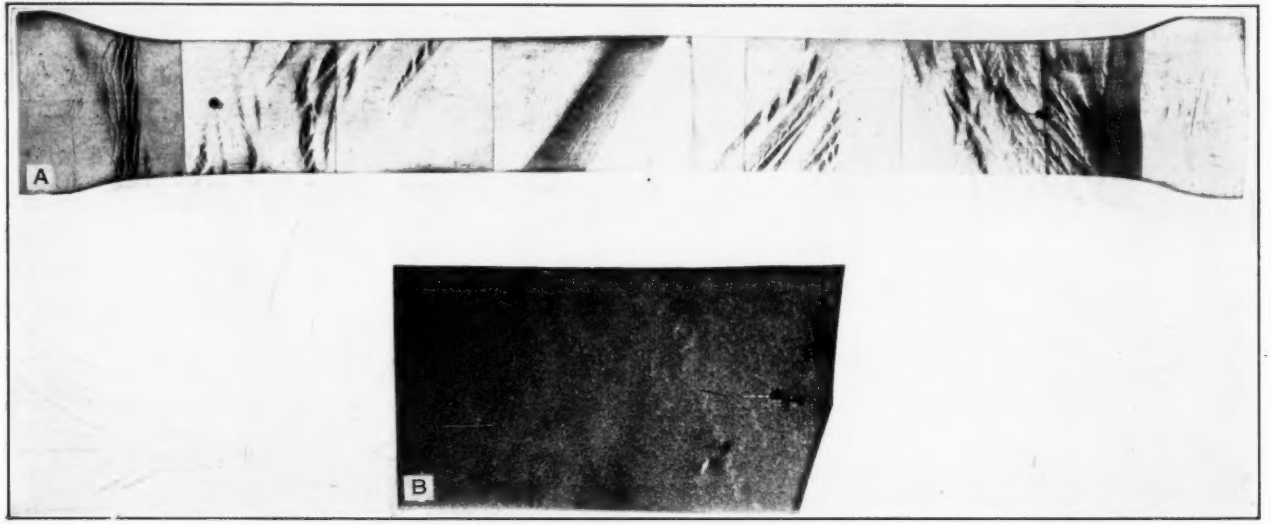

FIG. 5.—*Strain markings on specimen 7 (Table 1) after being loaded sufficiently to strain approximately the entire reduced section and also after being broken in tension*

a, appearance of the bar after loading 5, Figure 2. Note the diagonal section still free from strain markings, slightly less than × ½; *b*, one end (upper end of *a*) of specimen 7 after being broken in tension. × 1. Only traces of the most conspicuous preexisting strain markings persist

the fronts of the two deformation "waves" advancing from the ends of the specimen were not parallel. It will be seen from Figure 6 that although the two deformation "waves" had traveled nearly the length of the reduced section islands of apparently unstrained metal still remained at a considerable distance behind the "front."

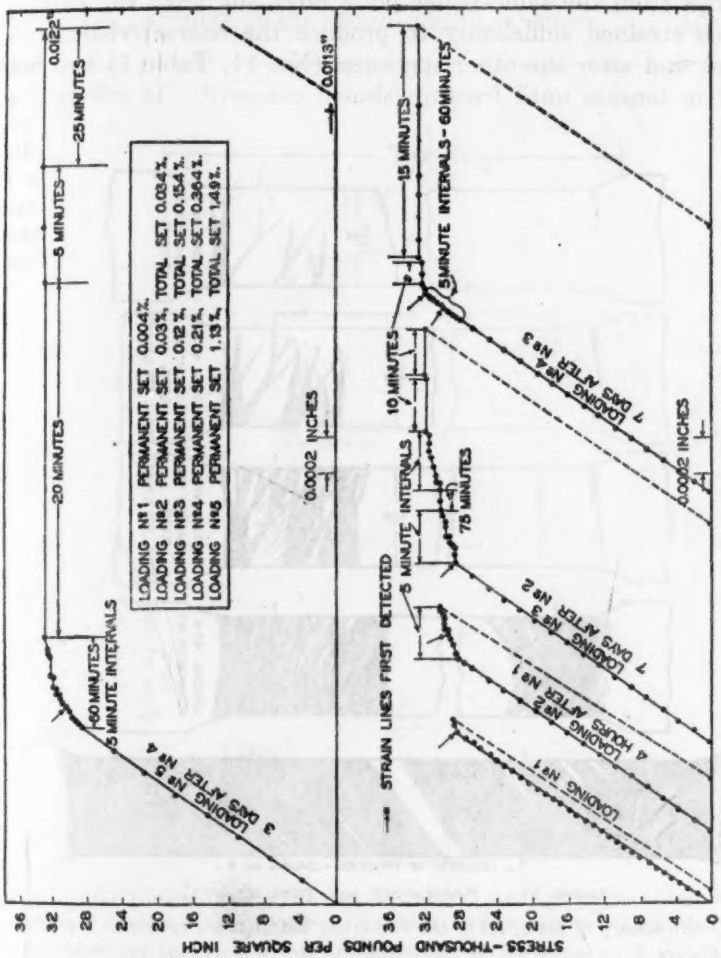

FIG. 2.—*Summary of the effects of the five successive loadings in which specimen 7 (Table 1) was subjected*

The results summarized in Figure 2 give no indication of strengthening of the material by overstraining. Since, as shown in Figure 6, the effect of each successive loading is confined to that portion of the reduced section which was not affected by the previous loading, the phenomenon of strengthening by overstraining would not be expected to occur until the entire reduced section had been affected; that is, had shown the "initial straining" effect.

3. HARDNESS OF STRAINED AREAS

It is well known that the effect of permanently straining mild steel is to increase its hardness. In Figure 7 are summarized the results of hardness tests by the Brinell and the Rockwell methods made on two of the 18-inch tension bars which were cut from adjacent strips from the same boiler plate after one (No. 10, Table 1) had been strained sufficiently to produce the characteristic strain markings and after the other specimen (No. 11, Table 1) had been stressed in tension until fracture almost occurred. It will be seen

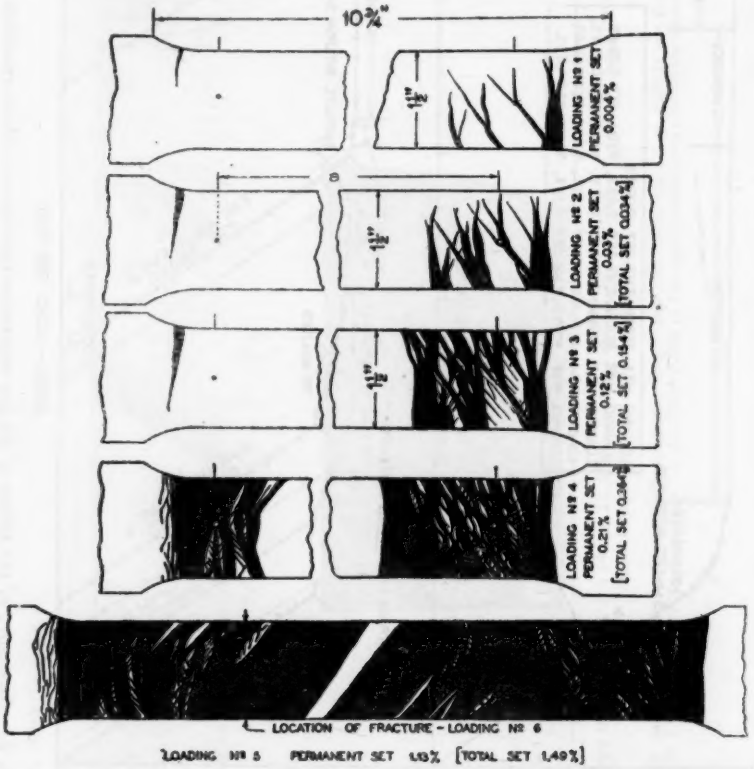

FIG. 6.—*Summary of the effect of the successive loadings on specimen 7 given in Figure 2, as shown by the corresponding strain markings produced*

that the hardness of the strained areas is somewhat greater than that of the adjacent "unstrained" metal which had the same hardness as the metal before the application of any load whatever.

The increase in hardness in the strain markings over that of the unstrained metal was slight, it varied in different specimens from 3 to 8 per cent according to the method used for measuring hardness. According to the Rockwell method an increase of hardness of 4.3 per cent was produced on one specimen (No. 3, Table 1). On the

same bar the Brinell method (10 mm ball, 3,000 kg load) indicated an increase in hardness of 3 per cent. On another bar (specimen 9) the effect of straining sufficiently to produce the strain figures caused an increase in hardness in the strain areas at the two ends of the bar of 3 and 5.5 per cent, respectively, as determined by the Brinell method in which a 1 mm ball and a load of 25.6 kg was used. The increase in hardness in the strain markings shown in Figure 7 over

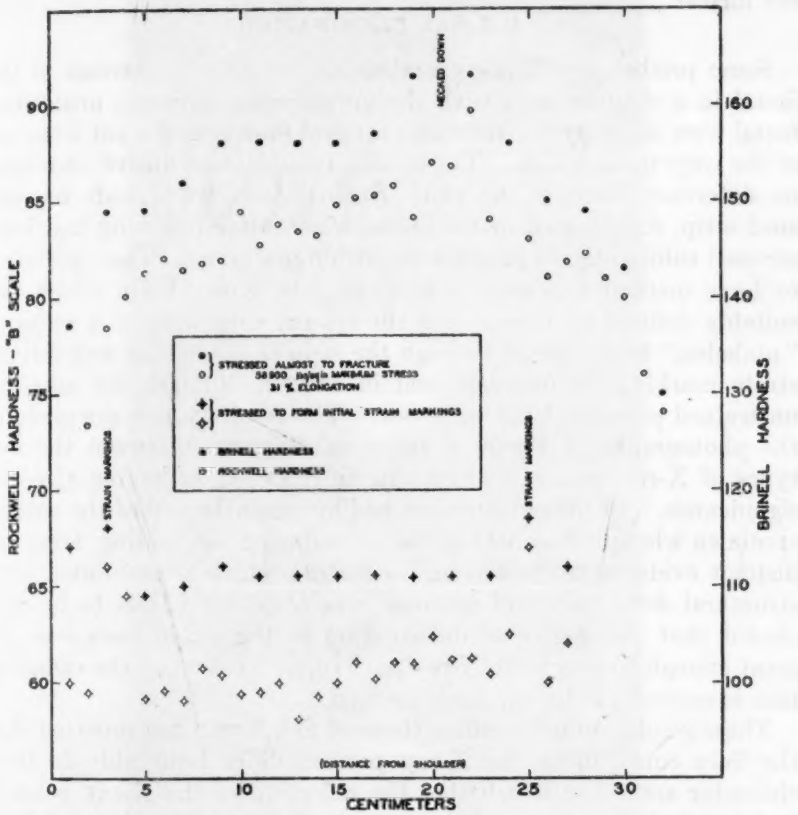

FIG. 7.—*Hardness survey of tension bars of soft steel (a) stressed sufficiently to form strain markings and (b) pulled until "necking down" almost to fracture occurred, specimens 10 and 11, respectively, Table 1*

the hardness of the unstrained metal was 5 per cent by the Brinell method and 8 per cent by the Rockwell.

These values are in fair agreement with the increase in hardness reported by Fell,[16] who found an increase of 6.8 per cent in the strain figures of soft steel produced by breaking a notched bar in tension. Fell's measurements, however, were made upon steel which had been tempered somewhat subsequently to straining, and this

[16] See footnote 8, p. 468.

fact may account for some of the difference between his results and those reported here.

The increase in hardness produced by the preliminary loading which resulted in the formation of the strain figures was small as compared with that produced by stretching the bar until fracture was produced. As shown in Figure 7 and Table 1, the maximum increase in hardness in the latter case was about ten times that of the former.

4. X-RAY EXAMINATION

Some preliminary X-ray examinations for the comparison of the metal in a strained area with the immediately adjacent unstrained metal were made by the reflection method on specimens cut from one of the large tension bars. The results, though inconclusive, indicated no difference between the two. Further tests were made on thin steel strip, 0.0025 inch in thickness, which after polishing had been stressed sufficiently to produce the strain markings. The "pinhole" or Laue method was used in this case, the X-ray beam which was suitably defined by means of a slit system consisting of a series of "pinholes," being passed through the strip in one of the well-defined strain markings in one case and in another through the adjacent unstrained portion of the same bar. The results which are given in the photographs of Figure 8 show no difference between the two types of X-ray patterns which can be regarded as having any real significance. In the pattern obtained by using the end of the tension specimen where it was held in the grips during the loading, however, distinct evidence of "asterism," a feature which is associated with structural deformation of a metal, was obtained. It is to be concluded that the degree of deformation in the strain lines was not great enough to affect the X-ray pattern, at least when the examination is carried out by the Laue method.

These results do not confirm those of Fell,[17] who has reported that the lines constituting the X-ray pattern differ noticeably in their character according to whether the reflection of the X-ray beam is from unstrained metal or from a strained area. The X-ray diffraction line shown by Fell as typical of the reflection from a strained area is described as "quite continuous, showing that the grain size is extremely small and/or that the crystals are distorted by work," whereas the line from the unstrained metal was "broken up, each dark spot being due to reflection from a small crystal."

It is apparent from other lines of evidence that deformation of the metal, undoubtedly by slip, occurs during loading and gives rise to the strain pattern. The X-ray method, which has proved so useful in other cases in explaining the "mechanism" of slip, is inapplicable

[17] See footnote 8, p. 468.

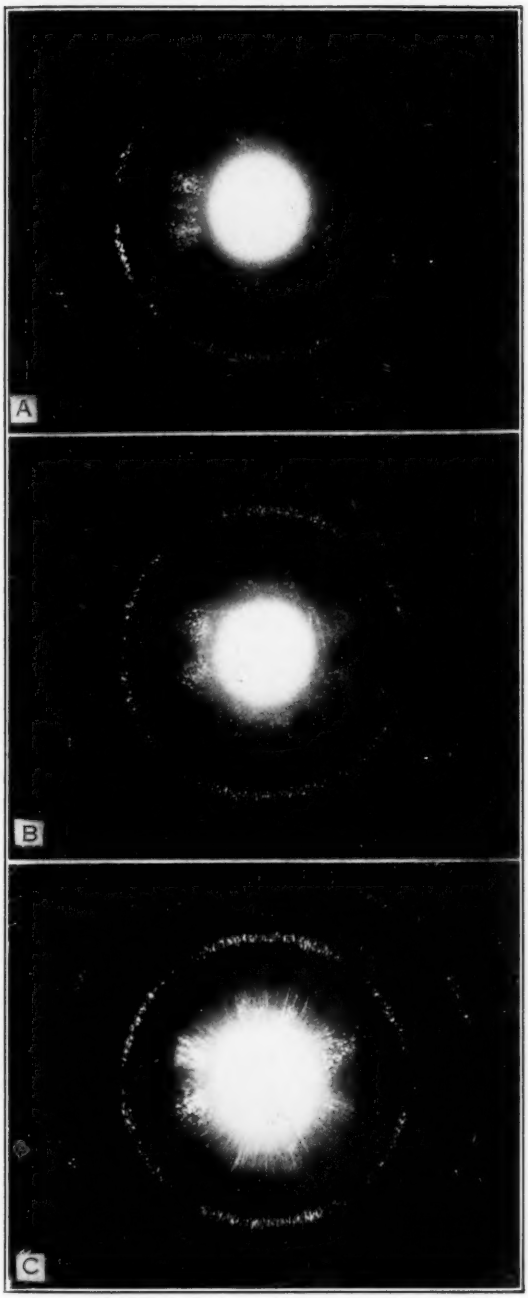

FIG. 8.—*X-ray patterns of low-carbon steel. Laue method. A specimen of annealed steel strip 0.0025 inch thick was stressed in tension until strain markings appeared*

a, "unstrained" metal adjacent to the strain markings; *b*, metal permanently strained, within one of the strain markings; *c*, end of the tension bar where it was gripped. Note the "asterism" caused by the severe deformation

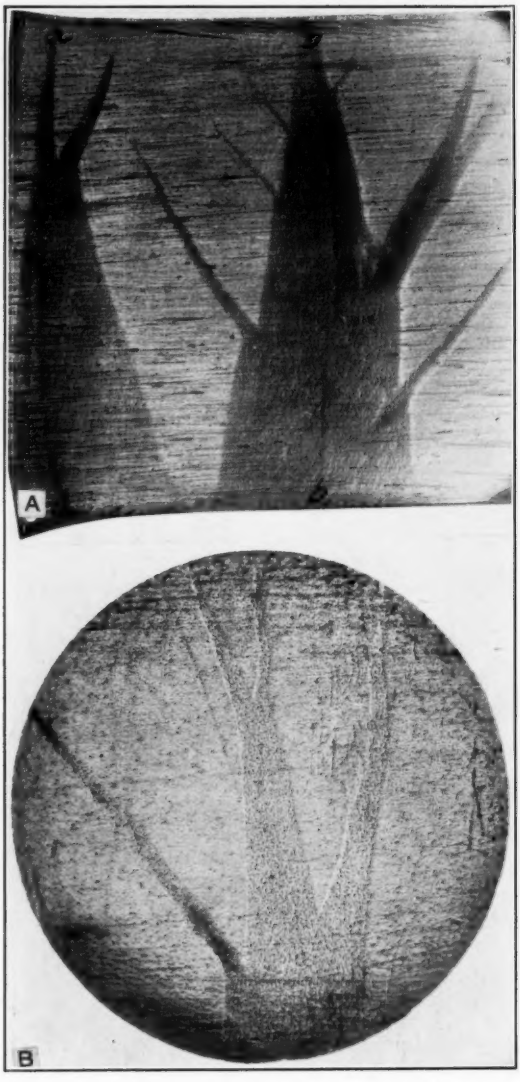

FIG. 9.—*Macroscopic appearance of the strain markings on low-carbon steel*

a, specimen 5 (Table 1). See Figure 11 (*a*). × 2. First line *a* formed at 28,600 lbs./in.², line *b* at 30,200 lbs./in.², load held for ten minutes and the wedges widened, the 45° branches formed at 32,000 lbs./in.², no flaking of the scale on the opposite side occurred; *b*, specimen 7 (Table 1). See Figure 3 (*b*). × 3½. Note the "rippled" appearance of the surface within the strain markings. The direction of stress application was parallel to the shorter dimension of the page

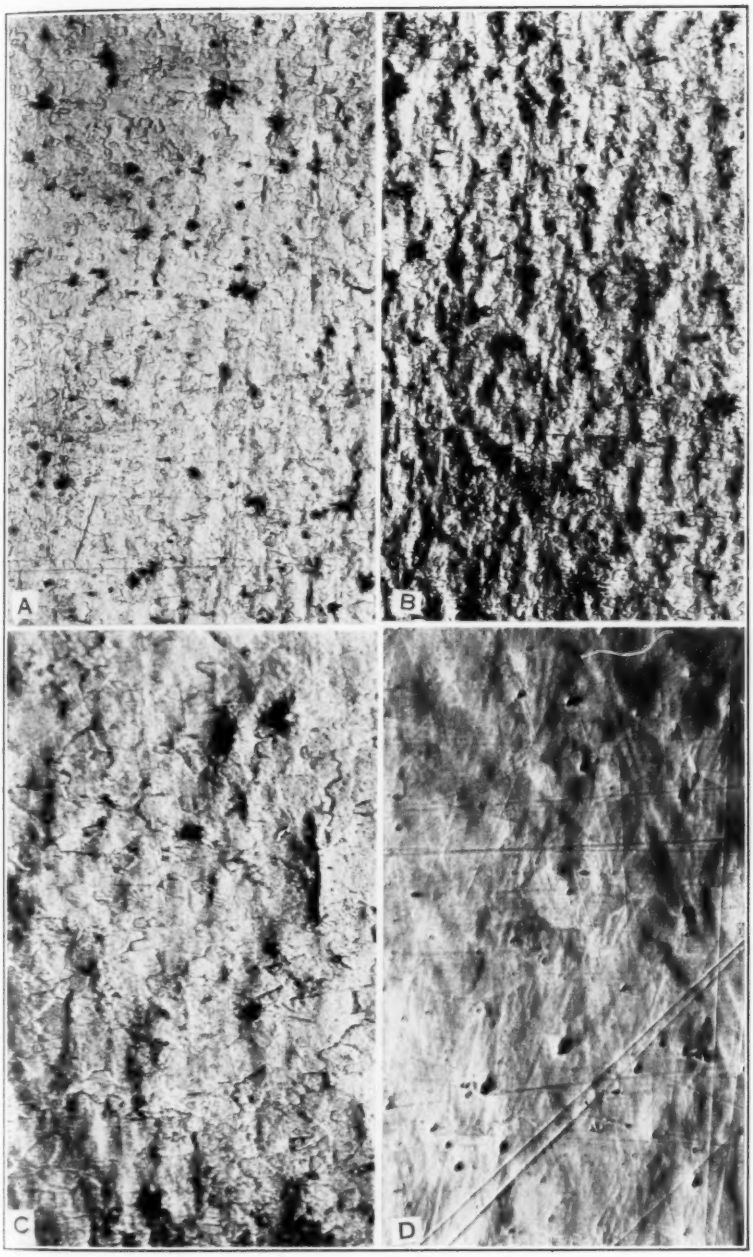

FIG. 10.—*Microscopic appearance of the surface of strain markings on low-carbon steel*

Polished strip steel, 0.0025 inch thick, was stressed in tension sufficiently to produce the strain markings, illumination, slightly oblique. *a*, The strained strip was etched to show the grain size. The micrograph shows the boundary of the strained area. × 100; *b*, similar to *a*, within a strain marking. × 100; *c*, similar to *b*. × 250; *d*, similar to *b*, unetched. × 250. The direction of stress application was parallel to the shorter dimension of the page. Note the relatively large size of the "ripples" in comparison to the ferrite grain size

UNIV. OF KY.

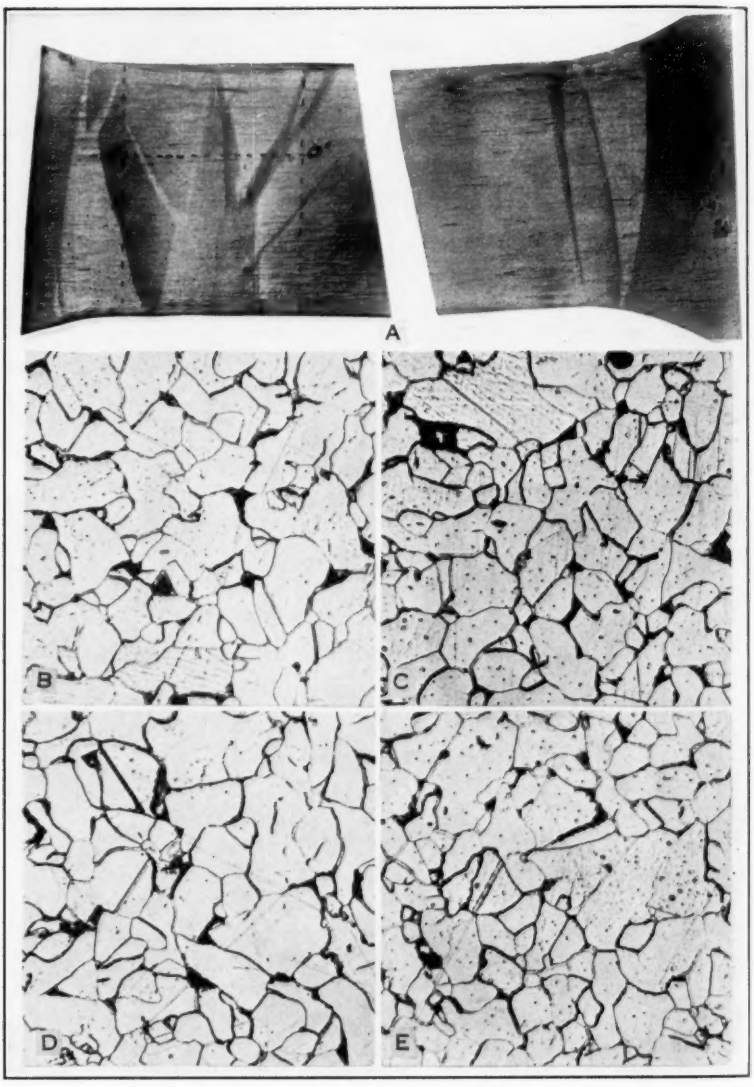

FIG. 11.—*Microstructure of low-carbon steel which has been strained sufficiently to form strain markings*

a, ends of specimen 5 (Table 1) showing strain markings. × 1. A section perpendicular to the surface of the bar along *o*—*o'* was prepared for microscopic examination and micrographs were assembled to show the structure of this entire strip; *b*, *c*, structure of the metal within the strain marking. × 250; *d*, *e*, structure of the metal outside of the strain marking. × 250. Etchant, picric acid solution. The structure within the strain marking does not differ essentially from that of the adjacent unstrained metal

here, however, and any attempt to correlate the strain effect with the crystal structure would be highly speculative.

5. MICROSTRUCTURE

When very slightly magnified, the strain markings are seen to have a "matt finish" in contrast to the polished surface out of which they have been formed. This is illustrated in Figure 9. A higher magnification reveals the fact that the matt appearance of the strain figures is the result of a great number of tiny "ripples," all of which are approximately parallel to one another and arranged perpendicularly to the direction in which the stress was applied. This feature is best seen with slightly oblique illumination and, as shown in Figure 9 (*b*), the boundary between the strained and the unstrained portions is a clear and definite one.

The size of these ripples on the surface of the strained areas is very much larger than that of the grains of the metal out of which they are formed. This is shown in Figure 10, which depicts the appearance of the surface of the 0.0025-inch steel strip which after polishing was etched so as to reveal the grain size and then stressed in tension sufficiently to produce the characteristic strain figures. The conclusion would seem to be warranted that in the production of these ripples many of the grains are shifted and tilted bodily possibly in addition to any internal distortion which they may undergo.

The examination of the structure of the interior of a strained bar at points corresponding to the surface strain markings failed to reveal any significant difference between the structure of the strained and the unstrained areas. This is in accord with the results of similar examinations reported by Fell,[18] who states:

> Attempts were made to study the difference in microstructure between the metal constituting the distortion wedge and the adjacent unstrained metal, but no sharply defined change was observed in the material in passing from the strained to the unstrained zones. Evidence of slip could be seen in both zones, although more frequently in the distortion wedges.

Figure 11 is representative of the results obtained in this study. A large number of micrographs were taken of a series of adjacent areas constituting a survey across the strain figure shown (specimen 5, Table 1). These were assembled in proper sequence into a long strip, over 10 feet long, for more careful study than was possible by visual examination alone with the miscroscope. The micrographs given in Figure 11 are representative. They show no marked features which may be considered as significant of the difference between the strained and the unstrained areas.

[18] See footnote 8, p. 468.

Specimens from the same tension bar showing prominent strain markings were annealed by heating to 650° C. and maintaining at this temperature for one hour, after which the microstructure was examined for evidence of recrystallization within the strained portions. As before, a considerable number of micrographs were taken which were assembled in proper sequence to show a "survey" of the structure across a strain marking. In Figure 12 are given micrographs representative of the metal of the strain marking and of the adjacent unstrained metal. As will be seen from the micrographs, both structures were essentially the same; the metal in the strain markings showed no greater evidence of recrystallization, occurring as a result of the annealing treatment which was given the steel, than did the unstrained metal.

A specimen of the 0.0025-inch strip steel was strained so as to produce conspicuous strain markings. The markings on the two sides of the thin strip were practically identical, and it is evident that the effect of straining as shown by the surface pattern extended entirely through the strip. The position of these markings was then definitely marked by sharp scratches along the boundaries. The surface was then polished for microscopic examination and etched. In Figure 13 are given micrographs representative of the metal within strain figures and of the adjacent unstrained metal. No difference in the microstructure of the metal in the two portions of the sheet, strained and unstrained, is shown by these micrographs.

A portion of the surface of a tension specimen of 0.0025-inch steel strip was polished, this being done by bending it into a loose coil and embedding a part of it in a matrix of sulphur which was poured around it while molten. It was then etched so as to reveal the grain boundaries and then, after removal from the sulphur matrix, stressed until strain markings showed. The microstructure of the surface corresponding to the strain markings and the adjacent unstrained metal is shown in Figures 14 and 15. The effect of the distortion upon the grain structure is unmistakable. It will be seen that many of the grains show evidence of distortion by slip, the slip bands, as revealed by their traces on the surface, being approximately parallel in the different grains and perpendicular to the direction in which the stress was applied. Not all of the grains show evidence of slip, and in those grains in which it has occurred the general shape and cross-sectional outline of the grain has not been noticeably affected. It is evident from these results that the formation of the strain markings is a slip phenomenon, the slip being confined to those grains which are suitably oriented with respect to the direction of applied stress.

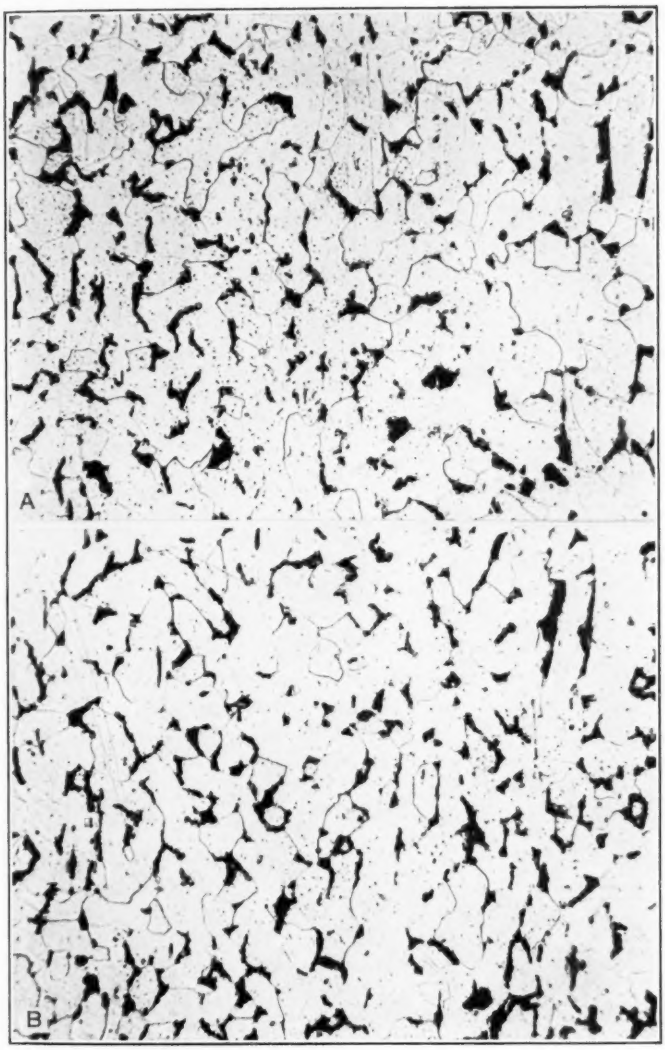

FIG. 12.—*Effect of annealing on the microstructure of steel showing strain markings.* × *250*

The material of Figure 11 was annealed one hour at 650° C. *a*, adjacent metal outside strain marking; *b*, metal within the strain marking. The straining which the steel received was not sufficient to cause it to show any greater tendency toward recrystallization than the adjacent unstrained metal

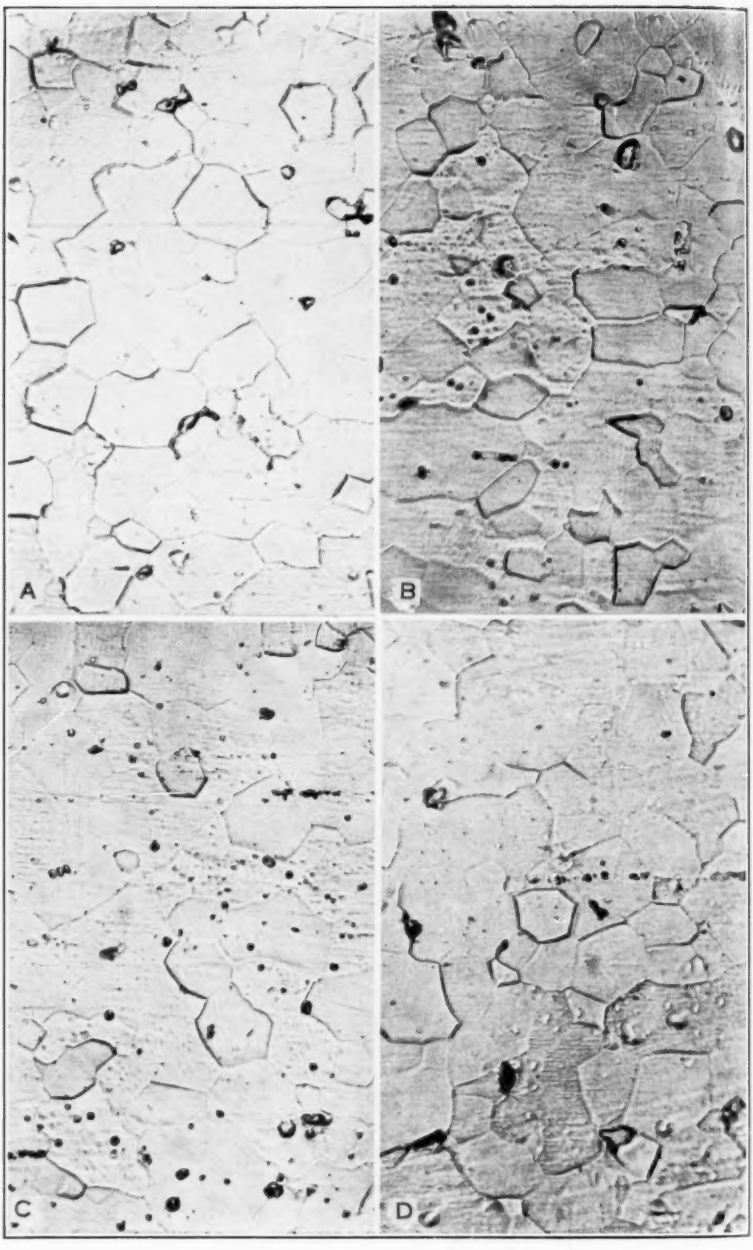

FIG. 13.—*Structure of low-carbon steel after straining.* × *500*

Strip steel (0.0025 inch) after being stressed so as to form strain markings was polished slightly and etched. *a* and *b*, structure of the metal outside of the strain markings; *c* and *d*, structure within the strain marking. No evidence of deformation in the structure is shown by this method of examination

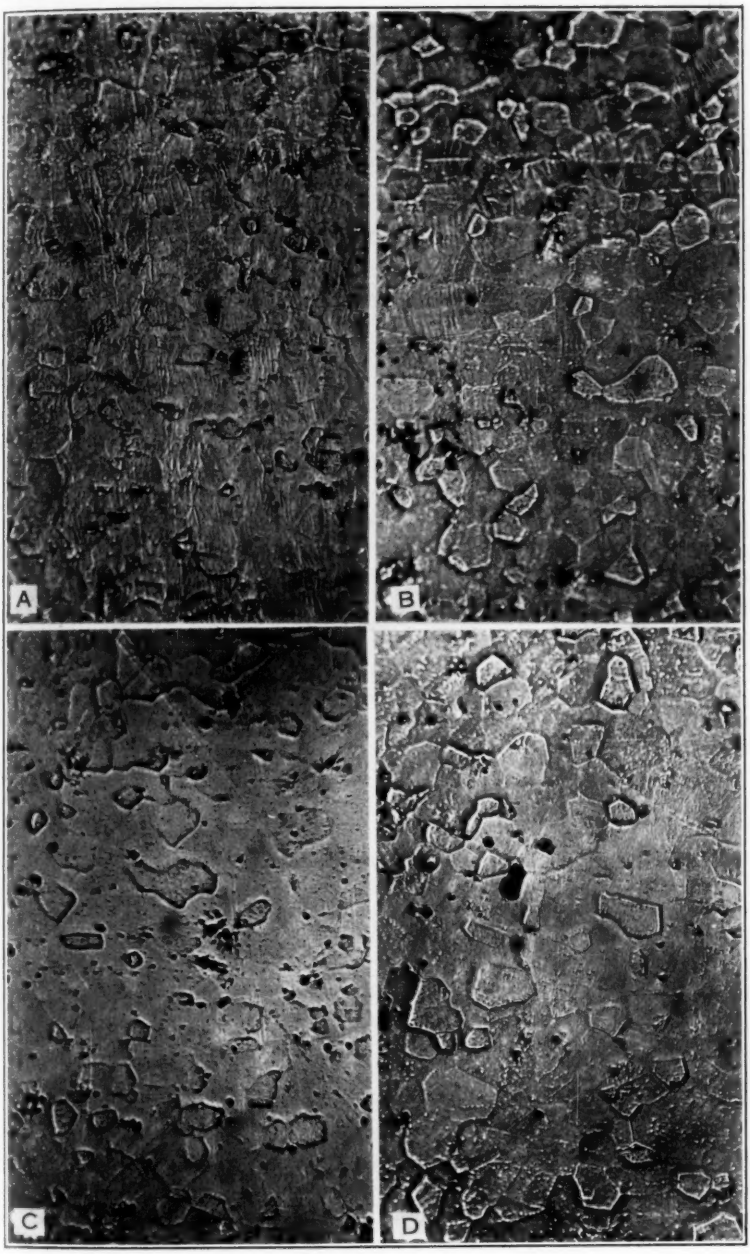

FIG. 14.—*Structure of low-carbon steel which has been stressed sufficiently to show strain markings.* × *250*

Strip steel after being polished and etched was stressed. *a* and *b*, structure within the strain markings; *c* and *d*, structure outside of the strain markings. Note the evidence of slip within some of the ferrite grains in the strain markings and the entire absence of it in the grains outside of the strain markings

UNIV. OF MICH.

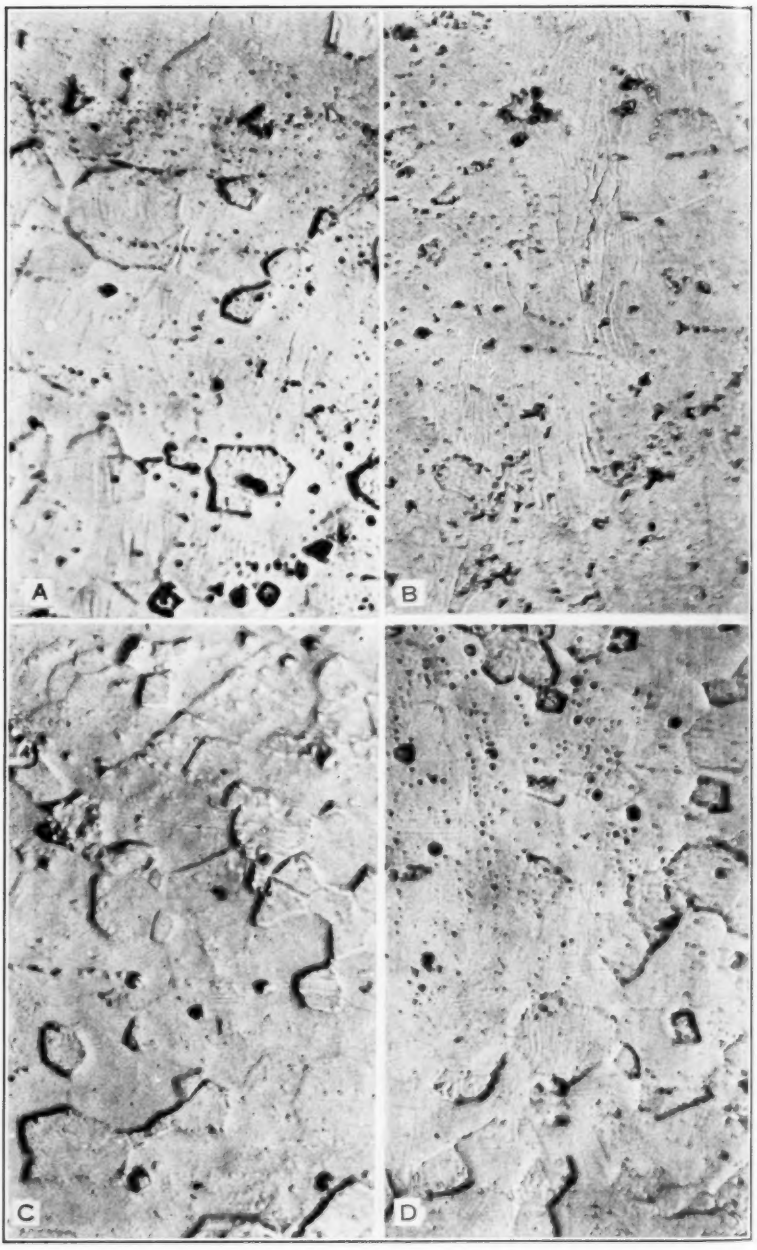

FIG. 15.—*Structure of low-carbon steel after being stressed sufficiently to show strain markings. Same material as in Figure 14.* × *500*

a and *b*, within the strain markings; *c*, outside of the strain markings; *d*, on the edge of the strain marking, left of the diagonal from lower left to upper right corner is outside the strain marking

6. COMPARISON WITH AN ALUMINUM ALLOY

A number of investigators have concluded that strain markings of the kind described in this report are peculiar to iron and soft steel. Jevons [19] suggests that the behavior of iron in this respect is a consequence of the body-centered cubic space lattice which is characteristic of iron. For purposes of comparison a few tests were made with an aluminum alloy sheet, aluminum being chosen because its crystal structure is different from that of iron and also because strain markings somewhat similar to those on soft steel have been reported [20] for such material. Tension specimens of the aluminum alloy, 25–S,[21] in sheet form, were stressed in tension. Figure 16 is representative of the results obtained. The stress markings, the formation of which is accompanied by an audible "click," appear as narrow bands extending entirely across the specimen at an angle of slightly less than 60° with the direction of the applied stress. The continued application of the load results in an increase in the number of such markings, the surface of the specimen becoming covered with a number of parallel bands. Unlike the corresponding strain markings on soft steel, no pronounced "growth" in the band width or in the formation of secondary "branches" was observed with the aluminum alloy under continued application of the load.

IV. DISCUSSION

Definite proof has been given in the foregoing sections that the markings on polished specimens of mild steel produced by straining, the so-called Piobert, Lüders', or Hartmann lines, are, as has generally been assumed, a special case of the phenomenon of deformation by slip. When a specimen, of the form of the standard tension bar, is subjected to tensile stress, strain markings first appear in the great majority of cases at the ends of the reduced portion of the bar close to the fillet—the most highly stressed portion of the bar according to photoelastic analysis. In the flat tension bars used in this investigation the first lines to appear were evidently of the nature of "alignment slip"—presumably a consequence of the lack of perfect alignment of the specimen with respect to the direction of the applied loading, even in the best "self-aligning" grips which were obtainable. These first markings usually appeared on the opposite sides of a flat

[19] See footnote 7, p. 468.

[20] R. J. Anderson, "Some mechanical properties of duralumin sheet as affected by heat treatment," Proc. Am. Soc. Test. Matls., **26** (2), p. 349; 1926; also, private communication from R. L. Templin, Aluminum Co. of America.

[21] Described in data sheet Al 5601, Handbook of Am. Soc. Steel Treat.; approximate composition of material used which was kindly supplied for these tests by R. L. Templin, Aluminum Co. of America: Cu, 4.2 per cent; Fe, 0.45 per cent; Si, 0.9 per cent; Mn, 0.68 per cent; Mg, 0.10 per cent. Aged at room temperature after being quenched in water from 520° C. (970° F.) approximately.

tensile bar, one at each of the two ends, and progressed across the bar nearly at right angles to the direction of the applied stress, as would be expected if this were an alignment effect. In general, as the load was maintained constant or only slightly increased, the first lines broadened into wedges as they progressed across the bar, although the position of the line of initial slip remained quite evident. Later, branches from the initial wedge appeared, which usually followed a course making an angle of about 50° with the axis of stress.

In general, as the load is maintained on the bar, the effect progresses toward the center of the bar; that is, away from the shoulder of the specimen by the formation of new markings originating slightly ahead of the advancing "front" in the form of lines which widen into wedges or bands at the edge of the bar. Often, however, "branches" extending relatively far into the unstrained metal in the form of narrow, pointed wedges formed from preexisting markings and constitute a conspicuous feature of the entire pattern. As the "growth" of the pattern progressed by the extension and widening of the lines, "islands" of apparently unstressed metal at very considerable distances back of the advancing "front" not infrequently were observed.

Within the strain markings deformation is not common to all the grains. Under the microscope only part of the grains were found to show evidence of slip, the slip bands in any one grain being nearly parallel to those in the other grains. Evidently only those grains which are so oriented with respect to the axis of stress as to permit slip at relatively low stress values, presumably along the planes of easiest slip, take part in the extension of the bar by slip.

The slight but noticeable roughening of the surface which gives rise to the strain markings is not caused solely by the slip within the favorably oriented grains. As a result of the slip that occurs "ripples" are formed on the surface, the size of which is much greater than the average size of the grains which constitute the metal. Evidently many of the grains in the immediate neighborhood of those in which slip occurs are tilted and shifted bodily to a slight extent, and give rise to the characteristic rippled surface appearance. Takaba and Okuda [22] have emphasized this aspect in their suggested explanation of the phenomenon. They have also stated that an essential condition for this phenomenon to occur is a sufficient number of grains, and that strain markings are rare in coarsely grained metal. These investigators, however, failed entirely to demonstrate the underlying fundamental cause—that is, deformation by slip within certain favorably oriented grains without which the effect could not be produced.

[22] See footnote 10, p. 468.

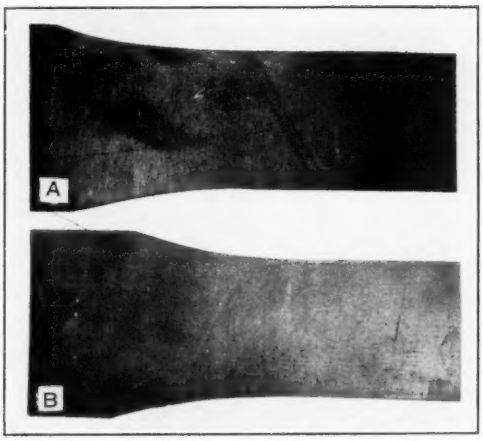

FIG. 16.—*Strain markings on an aluminum alloy sheet.* × *1*

The alloy, 25 S (see note 21) was used

It is very fascinating to watch a tension specimen while it is maintained under a load just sufficient to cause the formation of strain markings. The pronounced localization of the deformation; the very definite boundaries separating the permanently strained portions from the metal which, as yet, shows no evidence whatever of permanent strain; the slow, often intermittent, advance of the boundary as the condition of permanent strain advances into the "unstrained" metal, all serve to emphasize the fact that the process of the permanent straining or "yielding" of a metal bar in tension is progressive in nature. The permanent deformation of the entire bar is not accomplished instantaneously even after the first yielding has taken place. Observations on the behavior of a bar subjected to a series of successive loadings, each of which is just sufficient to cause yielding, serves to demonstrate this phenomenon very strikingly. When the bar was restressed, after having been removed from the machine in order that the effect of the previous load might be studied, it was seen that the straining process was resumed where it had "left off" when the previous load was released—that is, the strain markings which appear with the new loading form as an extension and continuation of those already present. Not until the straining effect has covered the entire reduced section of the specimen does there appear to be any further deformation in those portions which have already yielded under load. As was shown by the hardness determinations, these strained portions are hardened considerably by the cold-working they receive in the initial straining. Although the degree of strengthening of the steel is much less than that produced when the specimen is strained sufficiently to cause fracture, the steel is definitely stronger than the unstrained metal; hence, no further yielding would be expected in the initially strained portions until the stress on the bar had been increased somewhat.

As the load is gradually increased, after the initial straining of the entire reduced section has been accomplished, further extension takes place which is apparently quite uniformly distributed over the entire reduced section of the bar. This continues until the "necking-down" of the bar or localization of the deformation preparatory to fracture occurs. The pronounced flow which normally takes place in mild steel when loaded to the point of fracture wipes out almost completely all evidence of the previously formed strain markings. Only traces of the most conspicuous ones remain on the matt surface of the pulled bar, as is shown in Figure 5 (*b*). In the tension test as ordinarily carried out the rather rapid rate at which the load is applied, together with the fact that the surface of the specimen is not polished, render the effects described above, which constitute the yielding of the bar, inconspicuous so that they are often overlooked entirely.

It has been well established by the early work of Bauschinger,[23] Muir,[24] and others that the elastic properties of iron are very materially affected by overstraining the metal. It follows, however, from the results obtained by the progressive loading of a bar in tension, summarized in Figure 2, that the degree of overstrain to which a specimen—for example, a tension bar—is subjected must be sufficient so that the entire reduced section is permanently strained. Otherwise the elastic properties which will be measured when the load is applied a second time to the bar will be those of the initial material; that is, of those portions of the bar which were not permanently strained during the first loading. This follows directly from the fact that the permanent deformation of a bar under tension is decidedly localized. As already explained, only a small part of the bar is initially affected, and this effect progresses gradually along the length of the bar as the load is maintained. In the experiment summarized in Figures 2 and 6 during each of the four successive loadings after the first only a rather small additional part of the reduced section of the bar was permanently strained, the total effect of the five loadings constituting the straining of approximately the entire reduced section. The elastic properties measured during each of these five loadings, therefore, were those of that part of the reduced section of the bar which had escaped permanent straining in the previous loadings. Changes of the character of those reported by Bauschinger and Muir undoubtedly would occur within the strained portions. In a tension test, however, these changes are not apparent unless the entire reduced section of the bar has been subjected to overstrain.

Handford [25] has described a method for the measurement of slip in metals by means of which measurements of slip of the order of 65 Ångstrom units (1 Ångstrom unit = 10^{-8} cm) were reported. The method, which involves the use of thermionic tubes, depends upon the alteration of capacity of a condenser of which the specimen forms a part. The results reported for a normalized steel of approximately the same carbon content (0.25 per cent) as the material used in this investigation are of interest. The valve indicator showed slip at a stress of 31,800 lbs./in.^{2}, which is in very good agreement with the results obtained here.

A comparison of the tensile properties of single crystals of iron with the stress values at which the first evidence of permanent strain was observed is of interest. Edwards and Pfeil [26] have deter-

[23] J. Bauschinger, Über die Veränderung der Elasticitätsgrenze und Festigkeit des Eisens durch Strecken, Mittellungen aus dem Mechanisch-Technischen Laboratorium der K. Polytechnische Schule in München **13**, 1886.

[24] J. Muir, "On the recovery of iron from overstrain," Phil. Trans. Royal Soc. London, A 193, p. 1; 1900.

[25] C. Handford, "A valve method for detecting minute slip in metals," Phil. Mag. **47**, p. 896; 1924.

[26] C. A. Edwards and L. B. Pfeil, "The tensile properties of single crystals and the influence of crystal size upon the tensile properties of iron," J. Iron and Steel Inst., **112**, p. 79; 1925.

mined the tensile properties of iron crystals and reported values for the proportional limit varying from 4,250 to 5,500 lbs./in.^{2}, values for the tensile strength varying from 21,000 to 34,400 lbs./in.^{2}, and elongation 31 to 84.5 per cent. It is apparent that the behavior observed for any tensile specimen with respect to the stress at which deformation occurs is characteristic of the aggregate and bears no very obvious relationship to the individual grains which constitute the specimen.

The X-ray method of examination has given a great deal of information as to the mode of deformation occurring during the cold-working of metals in general. In the straining of metals just sufficiently to produce strain markings the total deformation is extremely slight as compared with that which occurs during ordinary cold-working, either by drawing or rolling. X-ray examinations, both by reflection and by the Laue method, have shown no marked difference between the unstrained metal and the same which has received the initial permanent strain resulting in surface strain markings. Microscopic examination has definitely shown evidence of slip in at least some of the grains. Any discussion, however, as to the probable crystallographic planes along which the slip occurs and the relationship of such planes to the direction of stress application would be highly speculative. The simplest assumption is that slip is confined to those grains in which the planes of "easy slip" are definitely oriented with respect to the 45° planes of maximum shear, there being, probably, an approximate parallelism between the two.

The term "yield point," as applied to metals, is usually defined as follows:[27] "The stress in a material at which there occurs a marked increase in strain without an increase in stress." It has generally been assumed that the appearance of strain markings of the kind described here indicates that the yield point of the material has been reached or slightly exceeded. That this assumption is sound is shown by the results obtained in this study. However, a definition of yield point based arbitrarily upon a given deformation for a given length—for example, 0.01-inch deformation over a length of 2 inches—may be open to criticism. If the results summarized in Figures 2 and 6 are examined on the basis of this arbitrary definition, the conclusion would seem to be justified that the specimen reached its "yield point" only after the fifth loading, although unmistakable evidence of strain markings existed after each of the four previous loadings. In fact, the determination of the stress at which the first evidence of distinct yielding is shown on a polished surface would seem to be a very good method for the determination of the proportional limit.

In carrying out this study of strain markings on steel only slight attention was given to the subject of revealing this effect by etching

[27] "Tentative methods of tension testing of metallic materials, E 8-24 T," Proc. Am. Soc. Test. Matls., 24, p. 1081; 1924.

reagents such as that described by Fry.[28] It was considered that the excellent results reported by Jevons[29] has fully demonstrated the correlation between strain effects and etch figures; further confirmation appeared to be unnecessary.

V. SUMMARY

1. A study of the strain markings, the so-called Lüders, Piobert, or Hartmann lines, on low-carbon steels has been made. Observations on polished specimens stressed in tension, together with strain measurements on the same, show that the first appearance of the strain markings on the polished surface coincided with the yielding of the metal and, in fact, indicated the proportional limit of the metal very closely.

2. The strain markings practically always appear first on a tension specimen at the ends of the reduced section, close to the fillet. The first lines to appear, in the work reported here, lacked the 45° orientation usually considered typical of these markings but formed approximately at right angles to the length of the bar. As the load is maintained or increased slightly above that at which yielding occurs, the strain markings gradually progress along the reduced section from the two ends. Extension of the strain pattern toward and into the shoulder ends of the specimen occurs only late in the tension test. The "propagation" of the strain along the bar shows itself by both the widening and branching of the lines and the formation of new ones usually just ahead of the advancing front. The examination of a specimen after being loaded several times in succession, with intermediate release of the load, shows that at each new loading the strain markings form as continuations and extensions of those already formed at the time of the release of the previous loading and continue in a similar manner with each new loading until the entire reduced section of the bar has been affected. No increase in the elastic properties by "overstrain" can occur until this has been accomplished. With further increase of the load, the strain markings are obscured as a result of the pronounced flow of the metal which, up to the time when "necking down" occurs, is distributed quite uniformly over the entire length of the reduced section. The location of the fracture in bars stressed to failure in the great majority of cases appears to bear no obvious relationship to the preexisting strain markings.

3. On surfaces covered with oxide (mill scale) strained areas can be located by the flaking of the scale. The method is a qualitative one, however, and can not be used for precise determination of the stress at which yielding occurs or for showing the fine details of the strain markings.

---

[28] See footnote 6, p. 468.

[29] See footnote 7, p. 468.

4. The hardness of the steel used in these tests was increased approximately 5 per cent locally by the straining which results in the formation of the strain markings. However, this increase in hardness is only about 10 per cent of the increase in hardness produced by straining the same material to the point of fracture.

5. X-ray examination failed to reveal any significant difference between the metal within the strain markings and the immediately adjacent metal which was not permanently strained.

6. Examination of the structure of the interior of the bar showed no significant difference in structure in the strained and the unstrained portions. After annealing at 650° C. for one hour no marked tendency toward recrystallization or grain growth was shown by the permanently strained metal. On the polished and etched surface of specimens subsequently strained sufficiently to produce strain markings evidence of slip was shown by some of the grains within the strain markings. The slip bands within any grain are approximately parallel to those within the other grains showing slip, and the traces of the slip planes as seen on the surface are nearly at right angles to the direction of the applied stress. Evidently slip occurs only within those ferrite grains which are favorably oriented with respect to the applied stress.

7. Under low magnifications the metal surface within the strain markings has a "rippled" appearance. Each individual ripple is very much larger than the ferrite grains constituting the metal. Evidently, as a result of the deformation by slip which occurs within some of the grains, other grains are tilted and shifted bodily, thus giving the previously polished surface a "matt finish" within the strain markings.

8. Observations on polished specimens serve admirably to demonstrate the progressive character of the yielding of a metal under tensile stress. A specimen may show in its strain marking very considerable evidence of having yielded and still be below its "yield point," according to some of the definitions of this term which depend upon an arbitrarily determined amount of extension over a given length.

9. Strain markings, similar in many respects to those which form on soft steel, were formed on an aluminum alloy sheet used for comparison.

The author is deeply indebted to J. R. Freeman, jr., metallurgist, for his interest in this work in conducting the tension tests and extensometer measurements, and to Samuel Epstein, associate metallurgist, for much of the photographic work. C. E. Eggenschwiler, junior scientific aid, has also contributed greatly in the attainment of the results, especially by his very careful work in the numerous microstructural studies necessary.

WASHINGTON, June 9, 1927.

○

The July, 1928, number (Vol. 1, No. 1) of the JOURNAL contains the following research papers:

RP1. Accelerated Tests of Organic Protective Coatings. Percy H. Walker and E. F. Hickson.

RP2. Measurement of the Tread Movement of Pneumatic Tires and a Discussion of the Probable Relation to Tread Wear. W. L. Holt and C. M. Cook.

RP3. Absolute Methods in Reflectometry. H. J. McNicholas.

RP4. Interferometer Measurements of Wave Lengths in the Vacuum Arc Spectra of Titanium and Other Elements. C. C. Kiess.

RP5. Analysis of Bauxite and of Refractories of High Alumina Content. G. E. F. Lundell and J. I. Hoffman.

The August, 1928, number (Vol. 1, No. 2) of the JOURNAL contains the following research papers:

RP6. Some Measurements of the Transmission of Ultra-Violet Radiation Through Various Kinds of Fabrics. W. W. Coblentz, R. Stair, and C. W. Schoffstall.

RP7. Tinting Strength of Pigments. H. D. Bruce.

RP8. Wave-Length Measurements in the Arc and Spark Spectra of Hafnium. William F. Meggers.

RP9. Tests of the Effect of Brackets in Reinforced Concrete Rigid Frames. Frank E. Richart.

RP10. Accelerated Laboratory Corrosion Test Methods for Zinc-Coated Steel. Edward C. Groesbeck and William A. Tucker.

These will be available as separates for purchase from the Superintendent of Documents, United States Government Printing Office, Washington, D. C.